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A STUDY ON
FRACTURE TOUGHNESS OF
ADVANCED STRUCTURAL COMPOSITES

by
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for

AIR VEHICLE TECHNOLOGY DEPARTMENT
NAVAL AIR DEVELOPMENT CENTER
WARMINSTER, PENNSYLVANIA 4

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Air Vehicle Technology Department

Naval Air Development Center

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FOREWORD

This report was prepared by Dr. George C. Chang under terms of Work Request No. WR2-0100, from the Air Vehicle Technology Department at the Naval Air Development Center, Warminster, Pennsylvania 18974. The report covers part-time research work performed during the period from 15 February 1972 to 15 July 1973. Mr. Anthony Manno, Code 30331, is the Project Engineer for the sponsoring agency.

ABSTRACT

The basic concept of structural fracture in homogeneous, isotropic materials is discussed. Fundamental relationships associated with the linear elastic fracture mechanics, based on the Griffith-Irwin approach, are presented. Discussed next are properties of fiber reinforced (or filamentary) composite materials from fracture mechanics point of view. Representative fracture test results are then reviewed and analyzed. Involved in the tests are numerous composites such as beryllium/aluminum, glass/epoxy, Scotchply, boron/aluminum, graphite/polyimide, etc. Some of these nonhomogeneous anisotropic composites are unidirectionally reinforced, while others are angle-plied. Finally, the fracture phenomenon and process in filamentary composites are discussed in details. It is found that the linear elastic fracture mechanics is applicable to most composite materials which have been investigated. It is nevertheless premature and risky to predict the applicability of the theory to the filamentary composite materials in general. Also made in this report are recommendations for future work on fracture of structural composites intended for high-performance flight vehicles.

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1. INTRODUCTION

Fracture in a structural material can be defined as an in-homogeneous process of deformation which causes regions of material to separate and load-carrying capacity to decrease to zero. This process can take place at varying speeds, depending on the nature of loading, the type of material, structural size, flaw size, etc. Some occurred rather rapidly, causing catastrophic failures.

For many years fracture has been found to occur even at stress levels well below the yield strength. Rashes of such brittle failures (in ships, airplanes and pressure vessels) have occurred with increasing frequency as the yield strength of materials have increased. As a result, the engineering community has had to give careful considerations to the fracture problem which may be present in a design.

The decade of 1960's saw the rapid development of advanced engineering materials, in particular, the filamentary composites. Advanced composite materials are combinations of high-strength, high-modulus filaments embedded in a (more or less) homogeneous, isotropic matrix. The matrix material may be a polymer resin, such as a relatively brittle epoxy, or a metal, such as the aluminum alloy. In a lamina, for example, the fibers contribute strength to

the composite, while the matrix maintains the orientation of fibers and transfers the loads between fibers. By properly orienting the desired amount of fibers in the selected matrix, the structural designer can obtain the mechanical properties required to support the demands that will be placed on the structure. Generally, the primary advantage a composite material offers is the relatively attractive strength-to-density ratio, and stiffness-to-density ratio. That is, most composite material can offer higher material strength than conventional isotropic materials, at a lower weight. This fits well into the recent aerospace requirements for high-performance hardware. Consequently, the composite material is on its way to become the most desired material within aerospace industry.

The history of the development of composites is a relatively short one. While some fundamental characteristics of the dozens of composites have been actively investigated, the science of <u>fracture</u> of filamentary composites remains rather primitive. The state-of-the-art in fracture toughness for composites is such that developmental tests must be conducted for each hardware article designed. It also points to the real need for a coordinated effort of years, involving both laboratory work and theoretical investigation, to expand the frontiers of knowledge in this important area.

The objective of the present work is to achieve a

realistic understanding of the phenomenon of fracture of structural composites. The ultimate goal of this effort is to develop analytical methods for determining fracture toughness as well as crack propagation characteristics of structural composites intended for aerospace applications. It is hoped that this modest effort, and its follow-on programs will contribute to further understanding of advanced composite materials in question.

2. FRACTURE OF CONVENTIONAL MATERIALS

The classical linear elastic fracture mechanics (LEFM) which deals with crack propagation in conventional homogeneous, isotropic materials, is based on the study of the stability of flaws, such as cuts, nicks, or voids. It is an outgrowth of the Griffith concept advanced in 1920. [26]*.

In the present section, a brief account of the theory will be discussed. This will serve as a stepstone leading into a discussion, in Sections 4 and 5 on the fracture behavior of the high-performance filamentary composites.

2.1. Fundamental Concepts of Fracture Mechanics

The original Griffith concept states that an existing crack will propagate in a cataclysmic fashion if the available elastic strain energy release rate exceeds the increase in surface energy of the crack. That is, the existing crack will propagate if thereby the total energy of the system is lowered. He stated that "the general conclusion may be drawn that the weakness of isotropic solids, as ordinarily met with, is due to the presence of disconti-

*Numbers in parentheses refer to entries in References.

nuities, or flaws, as they may be more correctly called, whose ruling dimensions are large compared with molecular distances. The effective strength of technical materials might increase ten or twenty times at least if these flaws can be eliminated." His theory gave a means for determining the correct relationship between fracture strength and size of flaws in a brittle material. Nevertheless, he did neglect the work in plastic deformation which is thought to occur at or near the crack-tip. Several authors have given useful discussions on the effect of neglecting the plasticity. [48], [51].

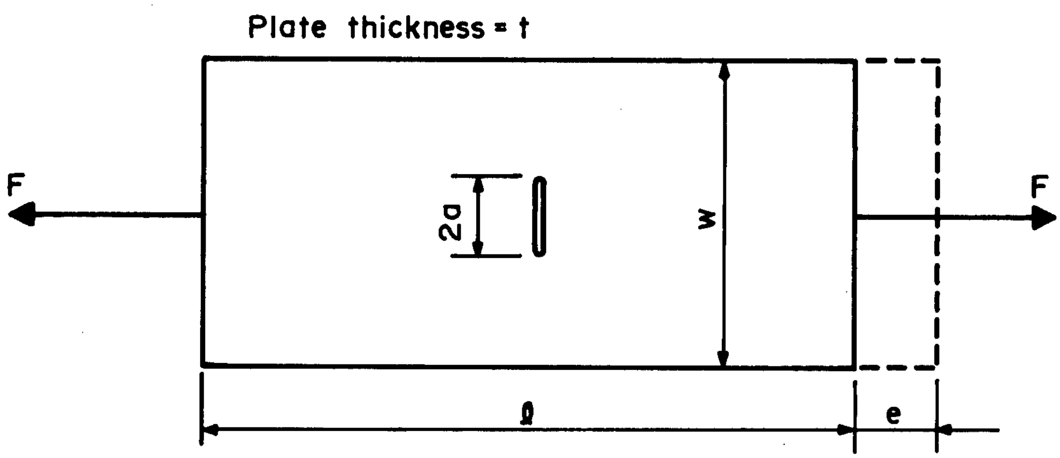
The question of plasticity, however, does not seem to be so important for filamentary composites. Most composites, such as boron-epoxy and the graphite-epoxy, have rather unique stress-strain curves exhibiting abrupt ruptures at ultimate stress level. Such experimental evidence indicates that these composites possess little plasticity, or strain hardening, which is characteristic of most metals, notably the mild steel. Thus, composites seem to be more brittle than metals.

In 1955, Irwin indicated that the energy approach is equivalent to a stress-intensity approach according to which fracture occurs when a critical stress distribution, characteristic of the material, is reached. An excellent account of the review on the equivalence of these two approaches is

given by Paris and Sih. [51].

2.2. Basic Relationships in Linear Elastic Fracture Mechanics

For convenience, the following sketch will be used in the present discussion concerning the basic mathematical relationships in fracture mechanics originally intended for homogenous isotropic materials.



The plate shown above has a crack with a lateral dimension 2a. A unit thickness (t=1 in.) is assumed. Upon loading to F_1, the elongation is equal to e. The stored energy at load F_1 is

$$\bar{E}_1 = \frac{1}{2}F_1 e$$

Consider an increment of crack extension, δa, resulting from a new loading, F_2. The stored energy in the plate after crack extension is, assuming the elongation e is maintained constant:

$$\bar{E}_2 = \frac{1}{2}F_2e$$

The increment $(\bar{E}_1 - \bar{E}_2)$ is then the released elastic strain energy associated with δa. Crack extension occurs only when $(\bar{E}_1 - \bar{E}_2)$ is equal to, or greater than, the surface energy required to form the new surface. Hence, the quantity $(\bar{E}_2 - \bar{E}_1)$ is, indirectly, a measure of the total work of crack extension over the increment δa.

Now, consider a definition of the release of elastic strain energy. The elastic energy stored in the plate is

$$\bar{E} = \frac{1}{2}Fe \tag{2.1}$$

Since the plate stiffness is

$$k = \frac{F}{e} \tag{2.2}$$

Eq. (2.1) can be rewritten

$$\bar{E} = \frac{1}{2}\,F\,\frac{F}{k} \tag{2.3}$$

Thus, the rate of change of stored elastic energy with respect to crack length, a, is

$$\frac{\partial \bar{E}}{\partial a} = \frac{1}{2}\,\frac{F}{k}\,\frac{\partial F}{\partial a} \tag{2.4}$$

Rewrite Eq. (2.2) as follows:

$$e = \frac{F}{k} \tag{2.4}$$

Differentiating above leads to:

$$\frac{\partial(\frac{F}{k})}{\partial a} = F\,\frac{\partial(\frac{1}{k})}{\partial a} + \frac{1}{k}\,\frac{\partial F}{\partial a} = \frac{\partial e}{\partial a} = 0 \tag{2.5}$$

where $\frac{\partial e}{\partial a} = 0$, since e is held constant.

It follows from Eqs. (2.4) and (2.5) that

$$\frac{\partial \bar{E}}{\partial a} = - \frac{1}{2} F^2 \frac{\partial (\frac{1}{K})}{\partial a} \tag{2.6}$$

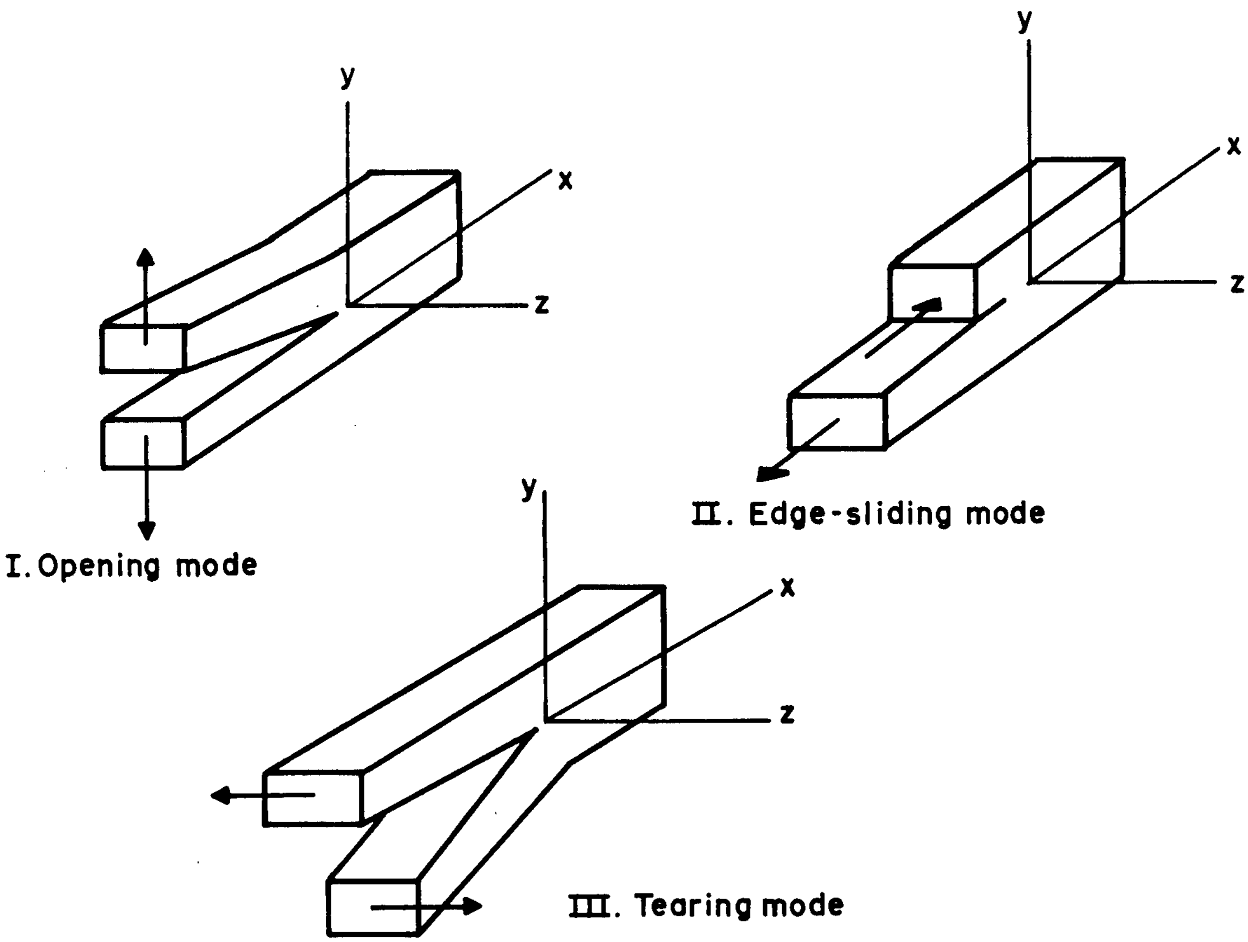
The rate of change of E is negative, indicates that energy is released. Defining G as the rate of release of elastic strain energy, it follows that

$$G = - \frac{\partial \bar{E}}{\partial a} = \frac{1}{2} F^2 \frac{\partial (\frac{1}{K})}{\partial a} \tag{2.7}$$

Referring to above equation, it can be said that the G value can be measured in a test. The load F can naturally be found in a test. The derivative term can be measured by compliance calibration. That is to say, it is the slope of the curve representing the relationship between $\frac{1}{K}$ and the crack length, a. Consequently, the rate of release of elastic strain energy of a fracture specimen may also be measured at fracture. As fracture toughness is defined as the work rate with respect to crack length to extend a crack, it is clear that fracture toughness is merely the critical value (upon fracture) of G. It is usually designated G_c.

Three modes of crack extension are possible. They are sketched on the next page.

The fracture toughness value associated with the opening mode, thus, is designated as G_{Ic}. Generally speaking, the three values for a given material bear no resemblance



to each other. They must be experimentally determined. In practice, G_{Ic} is often of interest.

As mentioned earlier, Irwin and others have shown that the fracture behavior of a material can also be characterized by using the stress intensity factor, K. These factors are different from the so-called stress concentration factors, although they are somewhat related. [32,53]. For the plane-stress condition, such as the plate considered earlier,

$$K_{Ic} = \sigma_f\sqrt{a} \qquad (2.8)$$

where σ_f is the nominal stress based on the gross area of the plate, i.e.,

$$\sigma_f = \frac{F}{1xw} \tag{2.9}$$

The above two equations remain valid when the crack length 2a does not exceed 0.3 xw, and when σ_f is below the yield strength of the material.

The relationship between the fracture toughness, G_{Ic}, and the stress intensity factor, K, has been well established. [23,51]. For plane stress

$$G_{Ic} = \frac{K_{Ic}^2}{E} \tag{2.10}$$

where E is the modulus of elasticity of the material. For the case of plane strain, the corresponding equations are

$$K_{Ic} = \sigma_f \sqrt{\pi a} \tag{2.11}$$

$$G_{Ic} = \frac{(1-\nu^2)}{E} K_{Ic}^2 \tag{2.12}$$

It should be noted that values for K for various cases with different combinations of loads and geometry can be found in several references [23,32,51,52].

Now the G-value can be determined in a laboratory test by loading a suitable specimen, as described earlier. Likewise, the K-value can be determined in a test. More details will be given in Section 4.

2.3. Some Facts about Fracture Toughness

It has been established that fracture toughness of a conventional material can be expressed in terms of either G or K. However, a number of facts that are of practical value will be discussed below.

2.3.1. Plane Strain vs Plane Stress

From Eqs. (2.8) and (2.10), it follows that the fracture toughness for plane stress is

$$G_{Ic} = \frac{1}{E} a \sigma_f^2 \tag{2.13}$$

Similarly, for plane strain, the fracture toughness is

$$G_{Ic} = \frac{(1-\nu^2)}{E} \pi a \sigma_f^2 \tag{2.14}$$

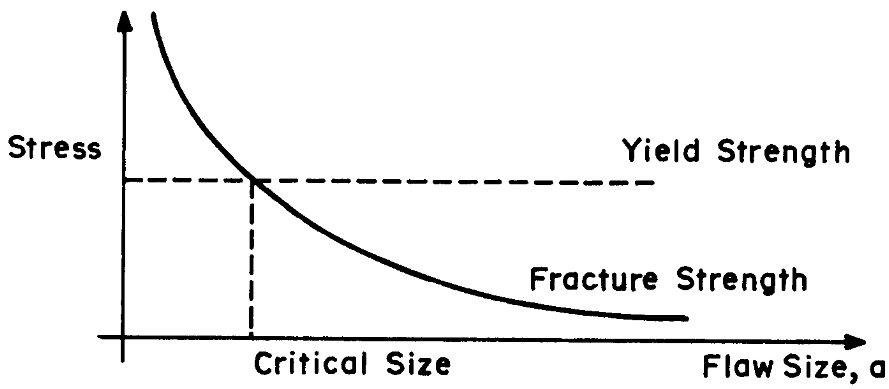
The two values differ by a healthy factor of $(1-\nu^2)\pi$. For steel, this works out to be roughly a factor of 3. Consequently, it is mandatory that the user be given the information as to the type of "fracture toughness" he is given.

Recent trend has been leaning towards the use of plane strain values. Several test programs of this nature will be presented in Section 4.

2.3.2. Material Strength vs Fracture Toughness

From Eq. (2.13), it is seen that the useful strength

of a material, based on σ_f, is inversely related to the flaw size, and directly proportional to the fracture toughness which is a material (physical) property. For a given material, therefore, the useful fracture-safe stress is solely dependent on the flaw size, a. The smaller the flaw size, the higher the fracture-safe stress, and vice versa.



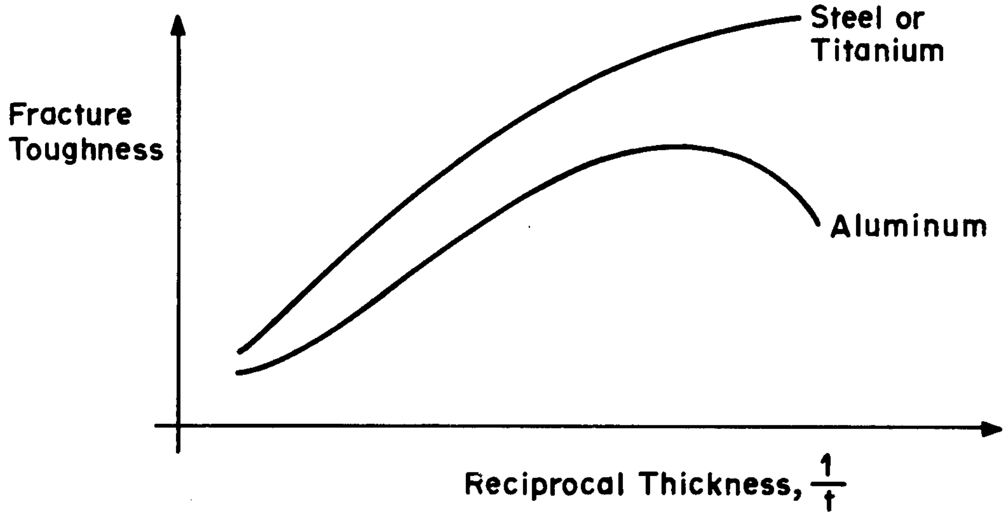
The sketch above clearly indicates that, for any material, there exists a critical flaw size corresponding to a given yield strength. Flaw sizes greater than the critical tend to reduce the usefulness of the material. In such cases, the yield strength can not be utilized as a datum based on which the safety factor is applied in a design. Here, the flaw dictates the design stress level.

2.3.3. Variation of Fracture Toughness

Evidence of significant variation in fracture toughness, of the order of 40 to 60%, of certain commercial

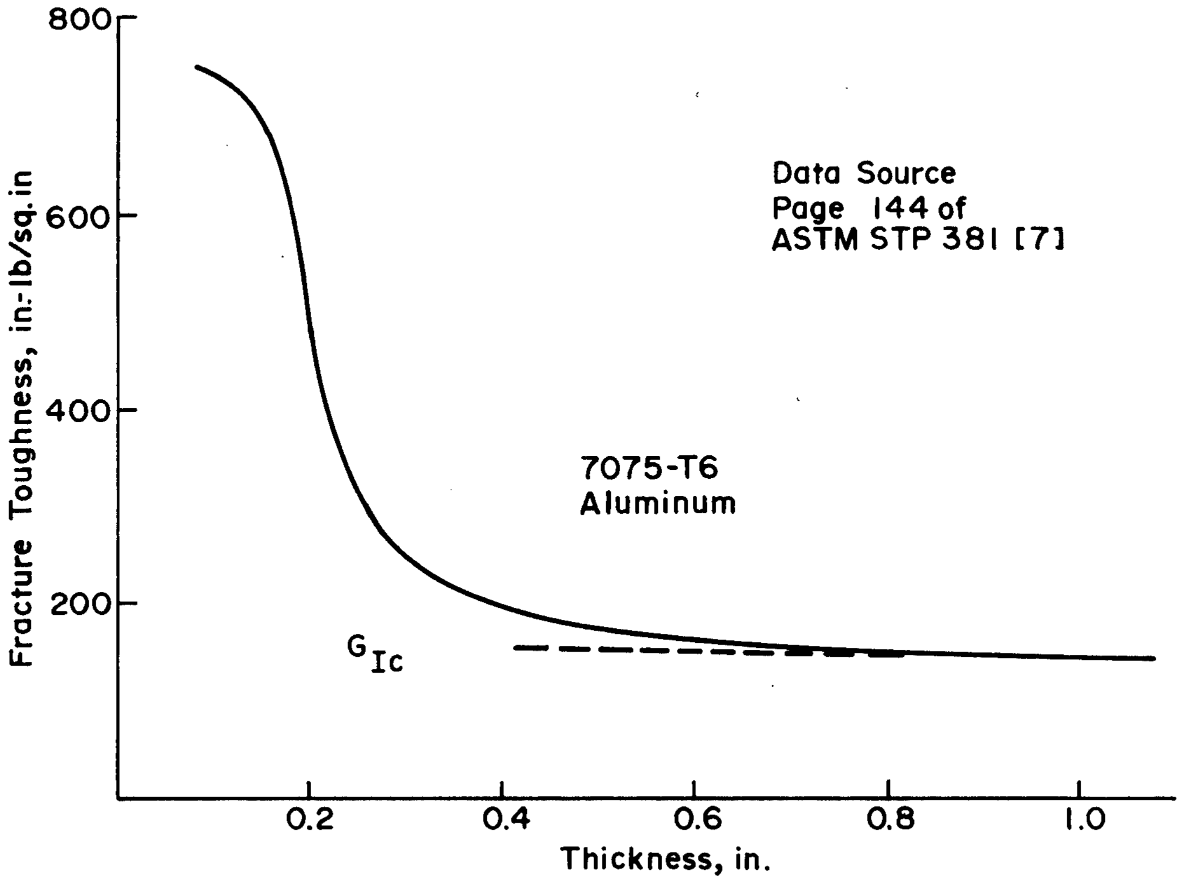
alloys has been reported. [52]. This is largely due to variation in chemical composition of the material. A second important contributing factor is the processing of the material, resulting in orthotropy as well as surface defects.

The thickness of material specimen also cast a very significant influence on fracture toughness. The general behavior is schematically given below: [69]



Thus, the thicker a material, the lower the fracture toughness, particularly for steel or titanium. The aluminum alloy shows some reversal though.

A more clear indication of the effect of thickness on fracture toughness is schematically shown in the curve below. It is based on data for 7075-T6 aluminum alloy. The curve is qualitatively typical of many high-strength metallic materials.

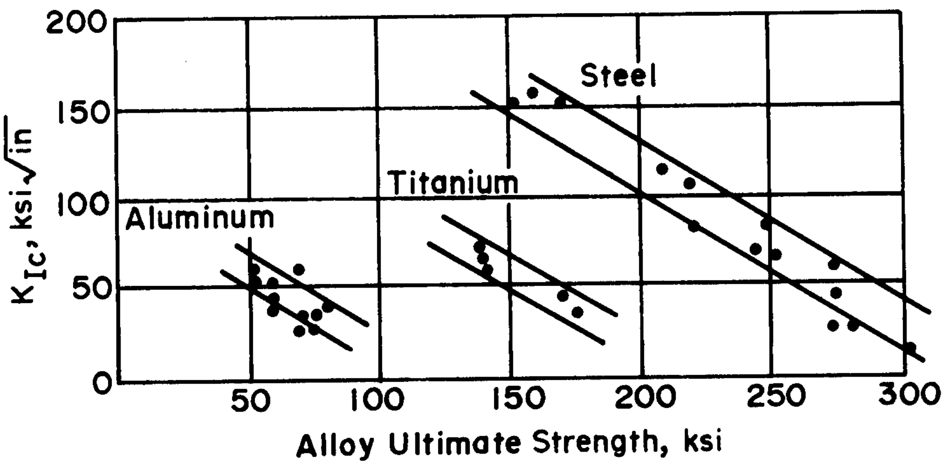


2.3.4. Fracture Toughness and Fatigue

Flaws invariably exist in most materials for engineering purposes. Under repeated loads, these flaws grow in size. Eventually, this type of crack propagation leads to complete failure of the whole member. Thus, the fatigue life of a machine component is also closely tied to the fracture toughness of, and flaws in, the material. A detailed discussion has been given by many authors. [45,62,64].

2.3.5. Fracture Toughness and Tensile Strength

Fracture toughness as a material property varies from one material to another. It also varies from one grade of a material to another grade of the same material. This fact is clearly illustrated in the following sketch. [64].



It should be noted that, for steel, the variations in K_{Ic} ranged from 35 to 155 ksi, representing a tremendous spread. A similar conclusion was detailed in a paper by Wei.[68]

2.3.6. Other Factors Affecting Material Fracture Toughness

Numerous other factors that affect the fracture toughness of a material exist. These include the temperature, chemical attacks (such as hydrogen embrittlement which is bad for high strength steels), stress corrosion, residual stresses, creep, etc. Temperature appears to have the most significant influence on the fracture behavior of a material. Most materials are said to have a ductile-to-brittle transition temperature. Thus, the fracture toughness measured

at room temperature when the material is semi-brittle, may become meaningless when it becomes brittle at a low temperature. Tetelman and McEvily gave a detailed account on these facts in their book. [62].

3. BEHAVIOR OF FILAMENTARY COMPOSITES

The decade of the sixties saw the most dynamic development in engineering materials, chiefly the fiber-reinforced plastics and its like. This type of material offers very high strength, yet weighs relatively little. As such, the filamentary composites promise to play an extremely important role in future aerospace vehicle construction.

In this section, relevant properties of composites will be discussed from fracture mechanics point of view.

3.1. The General Behavior of Composites

As is well known, a typical composite material is made of high-strength fibers embedded in a bonding matrix which has relatively low strength. The mechanical behavior of the composite, logically, depends on the corresponding behavior of the component materials. For instance, the graphite/epoxy composite system has been found to be different from even the boron/epoxy system, not to mention beryllium/aluminum, etc.

Since there are many fibers (glass, graphite, boron, steel, etc.) and a number of matrix materials available today, numerous combinations are possible. Consequently, it is believed that there exists a need to study the behavior of some of the desired composite systems on an individual basis. Results obtained in such a manner can be of engineering value. Unfortunately,

test results are scarce even on most composites under active investigation. This situation is particularly true with the study of fracture of composites. Therefore, in the course of this study, it is necessary to refer to test data from various kinds of composites. As a result, the observations as well as conclusions made herein are rather general, perhaps too general to be of value for any specific application.

Detailed discussions on the mechanical properties of various composites can be found in several books [4,5,6,9,25, and 47].

3.2. Composite Behavior and the Fracture Process

Several factors have very significant influence on the behavior of composites. They are discussed below.

Non-homogeneity and Anisotropy

One of the most pronounced characteristics of a composite material is its non-homogeneous, anisotropic behavior. As the fibers serve as reinforcements, this phenomenon is expected. The fibers are known to be extremely strong in tension longitudinally, and not so otherwise. Thus, the composite material becomes highly direction-dependent, i.e., being anisotropic. (The anisotropy can vary, though, according to the desire of the designer. It is an obvious advantage.)

As an immediate consequence, the strength of a composite varies. For example, the average tensile strength of a non-woven unidirectional laminate (continuous glass/epoxy, with 47 percent fibers by volume) was 160,000 psi. The same material in the cross-ply configuration was found to have a corresponding strength of only 75,000 psi. Its shear strength, however, is only of the order of 10,000 psi. Needless to say, the same can be true of the fracture behavior of composites. That is, fracture of filamentary composites can be highly direction-dependent.

High Strength

Generally speaking, composite materials have rather high tensile strength. This is certainly a desired feature from the viewpoint of specific strength. However, this high strength is not always realized, because of the fact that frequently the fracture strength is lower than the tensile strength.

In the case of conventional materials, it has been established that, for a given alloy material, the higher the strength the lower the fracture toughness. This fact is graphically illustrated in the figure presented in Section 2.3.5.

Now the composite materials generally have the strength of the steel as an upper limit. It is plausible that most

composites do follow the trend discussed above, but not necessarily have the same fracture toughness. That is, the K_{Ic} value decreases with increasing ultimate strength. However, this assertion must be verified by a series of tests.

Fiber Volume Ratio

It has been established that the strength characteristics of a composite material are influenced, to a very significant extent, by the amount of fiber present in a composite system [21]. The governing equation is:

$$\frac{\sigma_c}{\varepsilon_c} = E_f\, V_f + E_m\,(1-V_f)$$

Alternatively,

$$\frac{\text{Load in fibers}}{\text{Load in matrix}} = \frac{V_f}{1-V_f}\,\frac{E_f}{E_m}$$

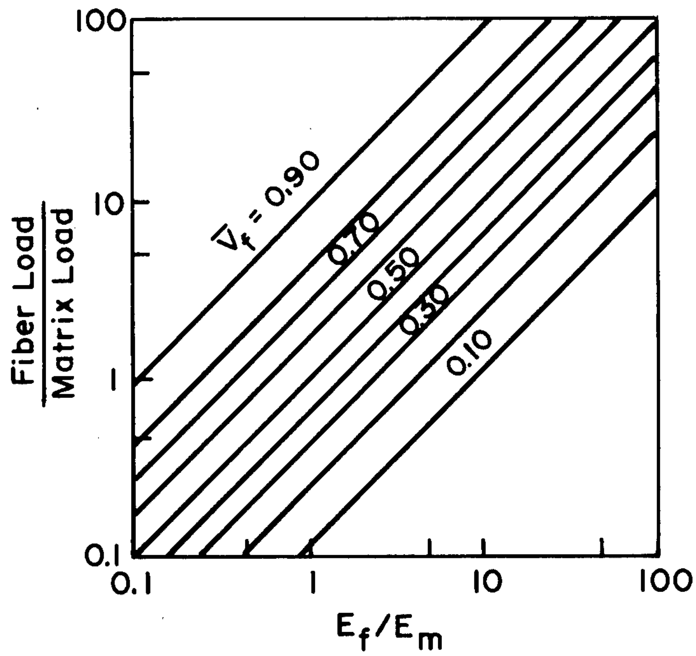
where V_f = fiber volume ratio

E_f = Young's modulus of fiber material

E_m = Young's modulus of matrix material

ε_c = strain in composite material

σ_c = stress in composite material



From the above, it can be seen that the distribution of load between the fiber and the matrix is heavily influenced by the fiber ratio. As a result of variation of loads of this type, the actual stress intensity factor (K) in a loaded composite material is also expected to vary.

Flaws and Manufacturing Quality

To this date, the quality of components made of composites has been largely determined by the workmanship. While some beneficial procedures have been developed, the results are still far from being satisfactory. Difficulties are still

encountered, even in the fabrication of laboratory specimens.

Very often, composite parts are found to contain flaws--voids or air bubbles, cuts, scratches and gouges. The distribution of fibers, most clearly shown in Fig. 4-3 as an example, is far from being ideal. All these factors attributed to manufacturing serve to weaken composite member. Moreover, the flaws are the worst, as they lead to significantly reduced fracture toughness.

Residual Stresses

The existence of residual stresses in composites is without question. Such stresses, when superimposed by the internal stress caused by external loading, further complicates the complex stress field in a composite member. For obvious reasons, the changed state of stress can result in a net reduction in fracture resistance. However, the state-of-the-art does not permit a quantitative assessment of the residual stress.

Length of Fiber

The length of fibers in a composite material is of significant importance. If the so-called discontinuous fibers which are shorter than a critical length are used, the composite strength usually is lower than that where "continuous" fibers are involved. This is caused by the fact that a portion of

each end of each discontinuous fiber is stressed at less than the maximum fiber stress of a continuous fiber. For example, the average composite strength for a 1/2-inch chopped fiber glass/epoxy composite (V_f = 0.48) was found to be 27,000 psi [21]. When compared with 160,000 psi for the nonwoven unidirectional laminate, this figure is indeed very low.

This appreciable reduction in composite strength naturally defeats the purpose of using such material. Nevertheless, the effect of such reduction on the fracture toughness is not clear at this time.

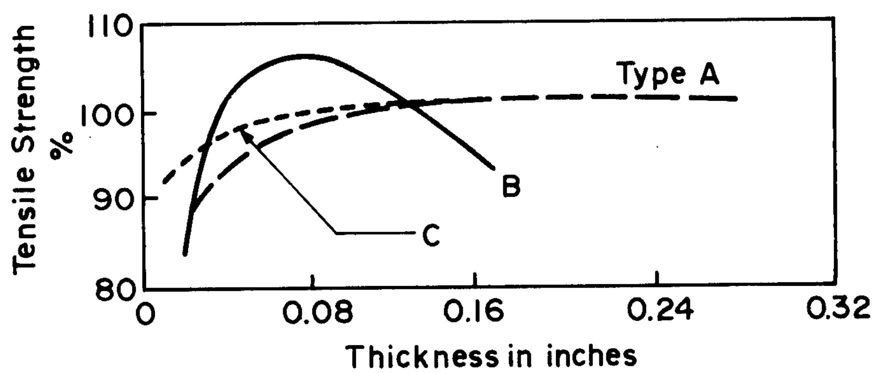
Bonding Strength

The bonding strength between the fiber and the matrix plays an important role in the strength of composites. The overall stress distribution is usually affected by the bonding properties. This change, in turn, can cause some variation in the fracture toughness of the material. A case in point is shown in Fig. 4-5, where crack branching and growth at interfaces in 6061-M-T6 aluminum-beryllium composite was observed [28].

Laminate Thickness

The thickness of a laminate influences the tensile, and

other strength. The tensile strength of thin laminates, 0.020 in. to 0.080 in., increases rapidly as thickness increases. The curves given in the figure below clearly indicate [40] that the effect of thickness becomes less profound for thicknesses greater than 0.08 in., although the evidence related to "Scotchply" is less than complete.



(A) Epon 828 CL, 112 Volan A (100% T.S. = 50,850 psi)

(B) Scotchply 1002, 181 Volan A (50,390 psi)

(C) Selectran 5003, 181-114 (40,050 psi)

The effect of thickness on fracture toughness of composites has not been clearly established. If the composites behave somewhat similar to that of isotropic material, then it is of practical interest to investigate this factor. In the case of isotropic material, the thickness of a test specimen should always be related to the plane stress or plane strain cases when toughness is to be determined.

Sensitivity to Environment

Composite materials are known to be sensitive to some environmental factors, such as moisture, and temperature. As these factors affect the composite strength, it is possible that they also influence the fracture toughness of the composite. The area is wide open with little literature in existence.

Other Factors

Additional factors which can influence fracture behavior of filamentary composites include: Modulus, ductility, and diameter of the fiber, modulus and ductility of the matrix. Furthermore, the selection of a matrix for a given fiber is of significant consequence. For example, the FORTAFIL 3-T graphite fibers were found to have a composite flexural strength of 200 ksi, when embedded in ERLA 4617/MDA Epoxy. The fiber volume ratio involved was 60%. What if the same fibers are embedded in a different matrix, say aluminum alloy 6061-T6? The outcome is obviously different, as the aluminum has a strength which is at least an order of magnitude higher than the epoxy. Since the same is also true of the fracture toughness ratio between these two matrices, it follows that the selection of a matrix will lead to a composite whose fracture behavior is significantly different from that involving a different matrix material.

Heat treatment appears to be another significant factor. Some experimental evidence of this contention is given in Section 4.2.2.

4. FRACTURE TEST OF FILAMENTARY COMPOSITES

Traditionally it is necessary to conduct many tests on a material before it can be utilized by engineers with confidence. The same, of course, applies to the fracture of composite materials in question.

4.1. Fracture Toughness Testing

For isotropic materials, there exists a huge volume of information on the testing of materials for the purpose of determining the fracture toughness. One particular book, the ASTM Special Technical Publication No. 381, contains a number of papers dealing with the subject matter. Its title reads: Fracture Toughness Testing and Its Applications. Theoretical as well as practical aspects of fracture toughness tests were discussed in these articles by competent authors. The topics range from specimen preparation and testing equipment requirements, to definition of G_{Ic} and K_{Ic}.

While a few articles in ASTM STP No. 381 were concerned with the testing of anisotropic materials, the bulk of the book dealt with isotropic materials. For the present purpose, it is somewhat appropriate to refer to the ASTM STP No. 460 entitled "Composite Materials: Testing and Design." A few publications outside No. 460 and 381 also appeared useful.

In particular, the ASTM STP 463 which came out in 1971 should prove useful for those interested in fracture toughness testing work.

It is noted with keen interest that many researchers have conducted composite fracture tests using the ASTM Tentative Test Method of Test E399 titled "Plane Strain Fracture Toughness of Metallic Materials". Details of the method are given in the 1971 Annual Book of ASTM Standards, Part 31.

The purpose of a fracture toughness test, of course, is to determine the G_{Ic} or K_{Ic} within the context of plane strain. It should be noted that the following relationship:

$$G_{Ic} = \frac{(1-\nu^2)}{E} K_{Ic}^2$$

which governs the isotropic material, is not necessarily meaningful for the composite material which is anisotropic. Briefly, it is so because of the fact that both ν and E are direction-dependent. (See Section 5.3 on G_{Ic} for composites.)

It is appropriate to mention here that the cost for a fracture toughness test can be very high. The specimens, particularly graphite/epoxy, etc., are very expensive. Glass/epoxy specimens, as verbally reported by several authors, are less costly. However, it is not so easy, if indeed meaningful, to extrapolate results obtained from glass/epoxy specimens to the graphite/epoxy. At this stage of development, only limited amount of work on glass/epoxy com-

posites has been performed. Work on graphite/epoxy is even more limited. Obviously, there exists a need to conduct carefully planned fracture toughness tests on the type of composite, such as graphite/epoxy, that shows promise for future growth.

4.2. Some Results of Fracture Testing

A review of published test results on the fracture of composite materials indicates that (static brittle or fatigue) fracture failure of such material almost invariably begins at a surface flaw (e.g., a scratch, nick, gouge, or cut.) [16, 17, 62,63]. Some tests, however, indicate that failure can originate in a weak fiber within the solid. [27]. As air bubbles or other flaws are known to exist within the composite, the possiblility of failure caused by such internal flaws cannot be ruled out at this time because of lack of experimental evidence, one way or the other.

A description of some selected testing work, as reported in recent literature, will be given below. No attempt is herein made to present results of all published work to date. A list of useful literature on subject matter is given in a bibliograph as a result of a computer search at DDC.

4.2.1. The Initiation and Growth of Fatigue Cracks in Filament-Reinforced Aluminum Alloys

The Test and Background Information

Reference [28] contained an account on a test-oriented project involving three different composites. The three were: beryllium/1235 aluminum, beryllium/6061 aluminum, and boron/7075 aluminum. The low-cycle fatigue crack behavior of these composites were compared. The effects of the strength and ductility of the filaments and matrices, and the role of interfaces on crack initiation and growth were evaluated.

Composite test specimens were fabricated by filament-winding and vacuum diffusion bonding. The beryllium/aluminum composite specimens were diffusion bonded for two hours at 475°C in a vacuum of 10^{-6} torr; the boron/aluminum composite specimens were bonded for one hour at 460°C in 10^{-6} torr.

The unidirectional beryllium/aluminum had a fiber volume ratio $V_f = 0.34$. Fiber size was 0.0050 in. ± 0.0001 in. as drawn (99.37 wt. % Be.).

The boron composite tested was based on a commercially produced high-purity version of 7075 aluminum alloy. The commercial boron fibers were 0.0040 in. ± 0.0001 in. in diameter. The unidirectionally reinforced composite had a fiber $V_f = 0.30$.

The composites were sectioned by electric-discharge

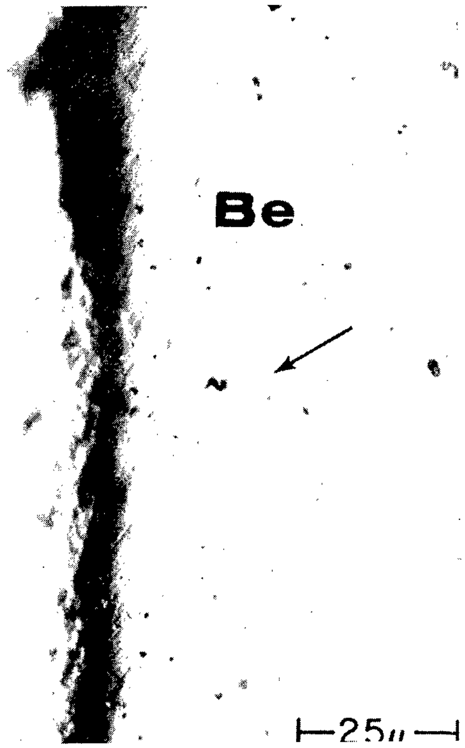
machining into test specimens. The approximate gage sections were 0.120 in. wide x 0.060 in. thick x 0.250 in. long.

Most tests were performed under zero-to-tension controlled axial strains applied parallel to the fibers.

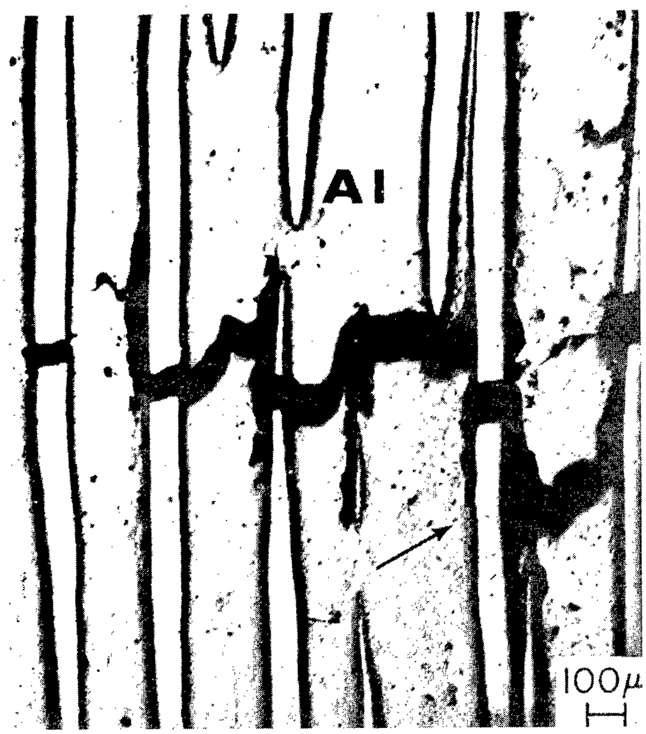
Metallurgical structure in the matrix, filaments, and interfacial regions were studied in an electron microscope. Fracture surfaces of composite specimens were studied in a scanning electron microscope. Longitudinal sections of partially failed composites were examined in a light microscope to determine fracture modes and crack-path morphology. Some of the observations made are shown in Figures 4-1 to 4-5.

Shown in Figure 4-1a is the fatigue crack initiated internally (arrow) in a beryllium filament. This indicates that a crack can develop within the composite material. In Figure 4-1b, the crack path in the beryllium/1235 aluminum is clearly shown. Both cracking (across the fibers) and splitting (parallel to fibers) took place. Figure 4-2 further exemplifies the composite fracture phenomenon. Figure 4-3 shows fractured surfaces in the beryllium/6061 aluminum. The boundary between fatigue fracture and tensile fracture is visible.

It was determined that the failure occurred through the progressive growth of a dominant fatigue crack across the



(a)
Crack Initiated in
Beryllium Filament



(b)
Crack Path in Be/Al
Composites

Figure 4-1 Fracture in Beryllium/Aluminum Composites

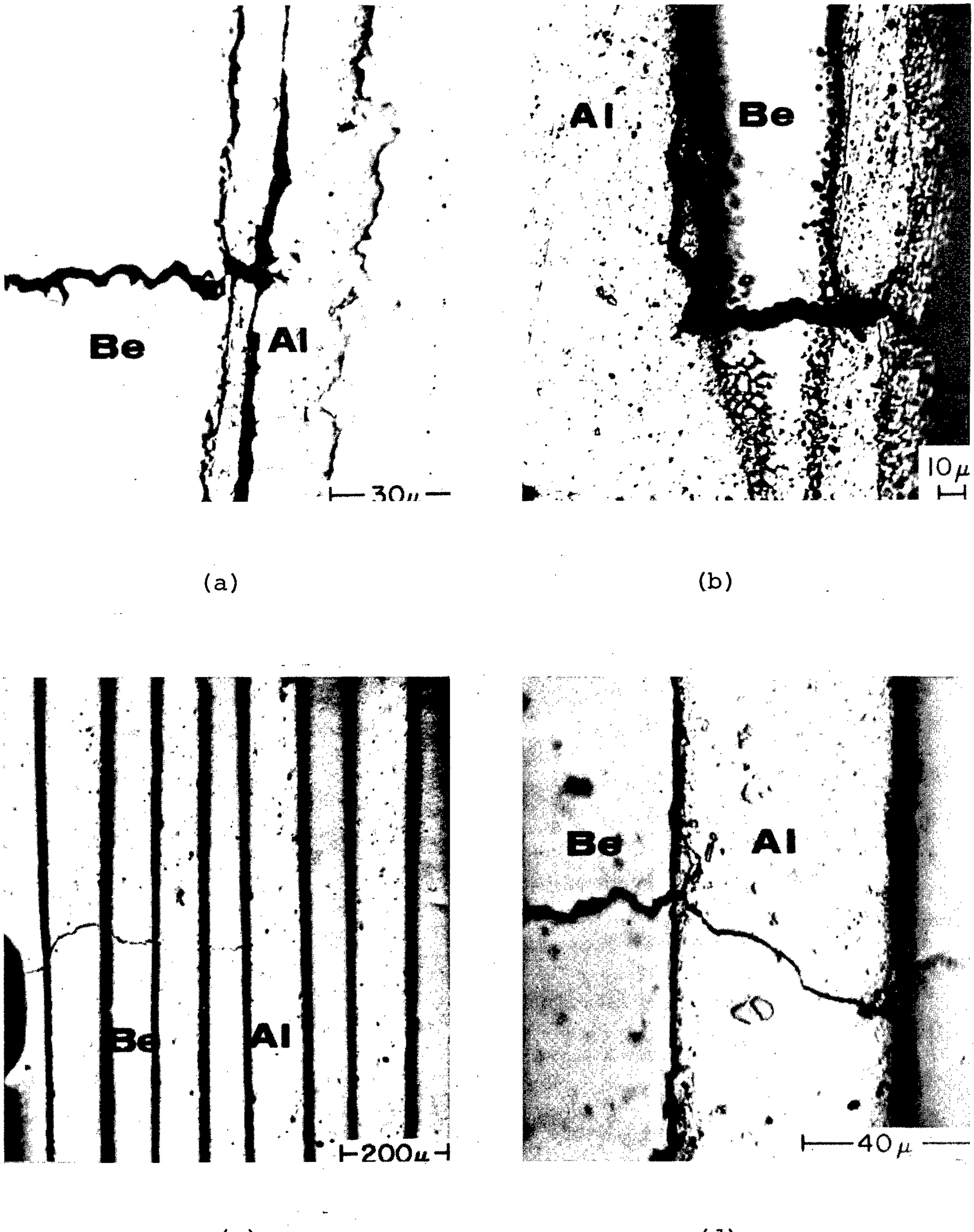
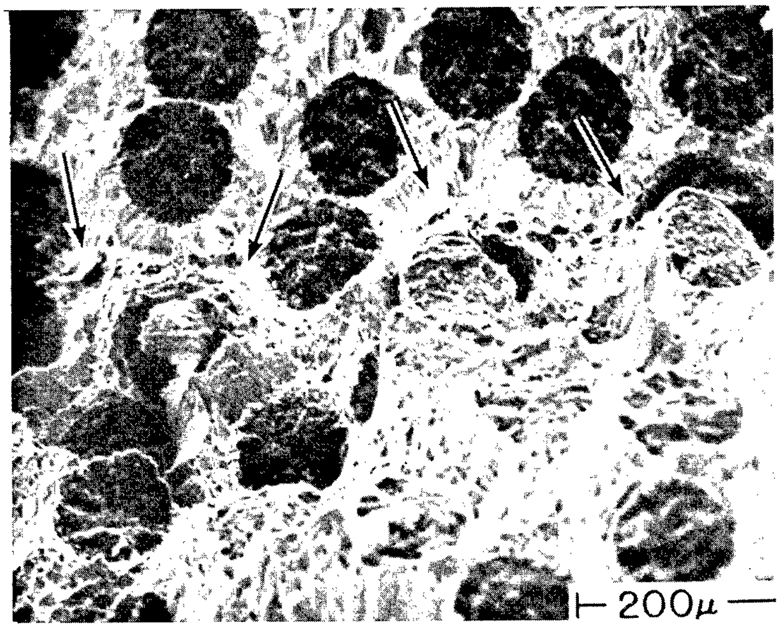
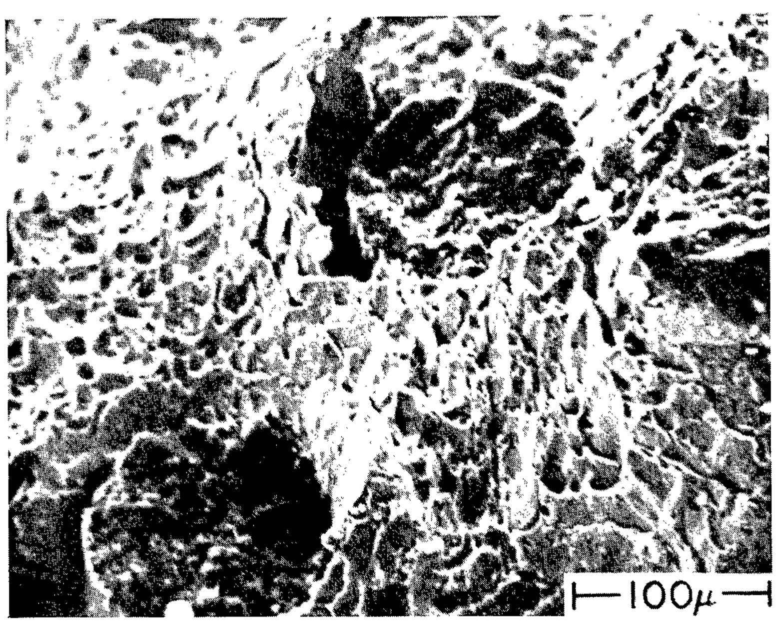


Figure 4-2 Cracks in Beryllium/Aluminum Composites

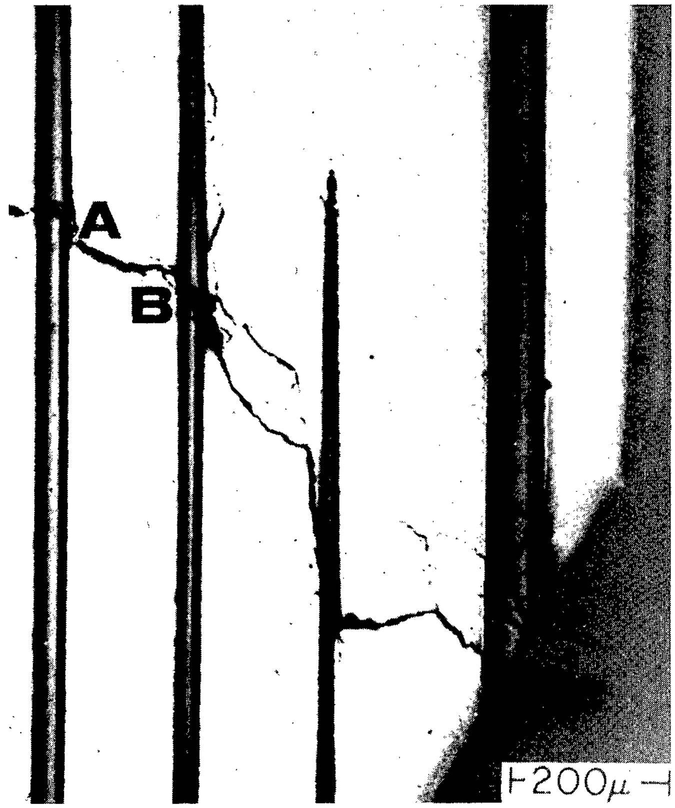


(a)

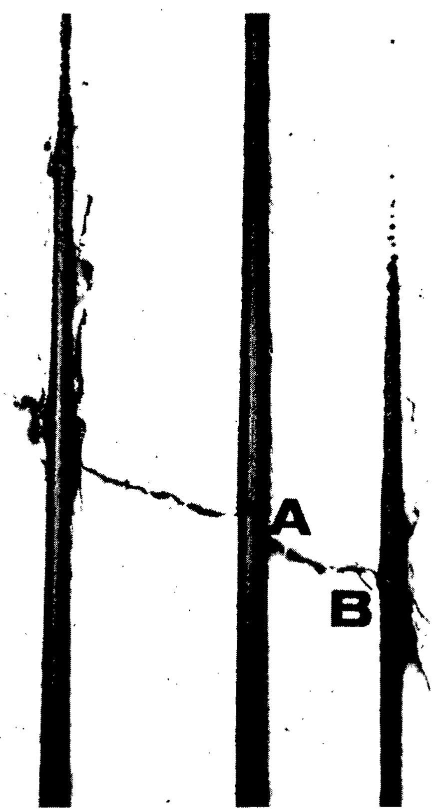


(b)

Figure 4-3 Fracture Surfaces in Beryllium/6061 Aluminum Composites



(b)



(a)

Figure 4-4 Cracks in Boron/7075 Aluminum Composites

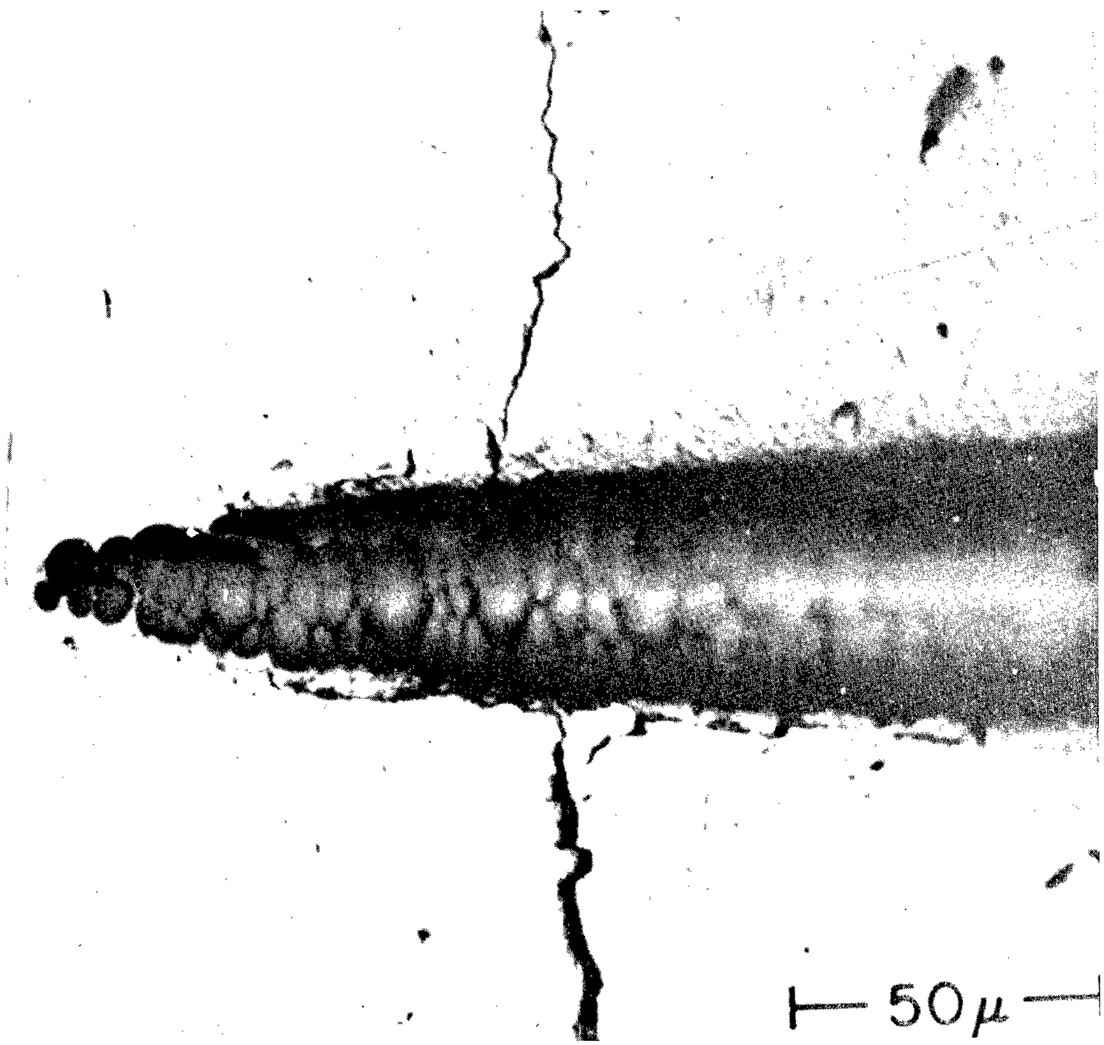


Figure 4-5 Fatigue-Crack Branching and Growth

specimen. The somewhat ductile beryllium filaments failed by the progressive growth of fatigue cracks which, however, did not initiate ahead of the main fatigue crack. The only apparent difference in the mechanisms of fatigue failure in this ductile composite system from that in homogeneous, isotropic alloys was the effect of interfaces on growth of fatigue cracks. Here, weak interfaces and a low-yield-stress matrix promoted crack branching at interfaces, and were somewhat effective in inhibiting the transverse growth of fatigue cracks.

This impeding effect is also exhibited by the boron/7075 aluminum system. Figure 4-5 illustrates the branching mode and crack growth along the filament in a mode II (shear component of stress applied normal to the leading edge of the crack) type of crack growth; apparently, cracks also circumvented filaments in mode III (antiplane-strain) crack growth. Several broken boron filaments were found randomly located in this composite system after the test. It is not clear whether this phenomenon corresponds to Toth's postulate that fatigue crack can initiate internally. [65].

As a result of this test, it was concluded that (1) the relative magnitude of the principal elastic stresses around the crack tip are very sensitive to the ratio of elastic moduli of the two materials, and (2) the shear and tensile strengths of the interfaces exceeded those of the matrix.

Furthermore, it was discovered that ductile beryllium filaments failed by the progressive growth of fatigue cracks, whereas brittle boron filaments failed suddenly. As a result, the fatigue resistance of beryllium filaments is very sensitive to the magnitude of crack-tip stresses, and hence to matrix yield strenth, while boron filaments are virtually insensitive to crack-tip stresses below their fracture stress. Ductile filaments are clearly less effective than brittle filaments as crack stoppers. Thus, maximum resistance to the growth of fatigue cracks is expected for a composite system employing brittle filaments and a low yield-strength matrix.

Discussion

A question might arise at this point, just why this work is reviewed in such great details? The answer is twofold. First, this document contains a number of valuable photos such as those presented herein above. Being a government publication, the contents can be reproduced. Such is usually not true with most other publications. Secondly, the fatigue fracture reported here bears a very close resemblance to most "static" brittle fracture, as far as crack growth and failure modes are concerned. The initiation of cracks, of course, marks the difference between the two. This second point can be substantiated by reviewing several related documents. [11,18, 29, 33, 34,37.]

4.2.2. Toughness of Filamentary Boron-Aluminum Composites

The Test and Background Information

Hancock and Swanson [29] reported an exploratory study concerning the fracture toughness of filamentary composites. The study was carried out to determine whether valid fracture toughness values could be measured for boron aluminum composites.

Commercial 6061-T6 aluminum alloy was unidirectionally reinforced with approximately 30%, by volume, of continuous boron filaments of 0.0040 in. diameter with tungsten core (Source: The Hamilton Standard Division of United Aircraft Corporation, Windsor Locks, Connecticut). Composite specimens of three different thicknesses (0.032 in., 0.073 in, and 0.100 in.) were fabricated by filament-winding and vacuum-diffusion-bonding methods. Diffusion bonding was carried out at 475°C for 1 hour, at about 10^{-6} torr. The specimens were subsequently sectioned and center-notched by electric discharge machining (EDM). The specimens had a width of approximately 1 in., with a center notch length of about 0.26 in. (=2a). The average notch-tip radius was estimated to be approximately 0.001 in. These specimens were heat-treated to the T6 condition or to a modified (M) T6 condition, which differs from the T6 only by cooling the sample to -196°C for 4 minutes just prior to aging.

These specimens were tested on an Instron testing machine (Model TTCM) at a cross-head of 0.02 inch per minute, simulating a "static" tensile test. The stress-intensity factors were then calculated from the K-calibration expression for center-notched metal plate [10], assuming it is also applicable to composites:

$$K_I = 1.77\ (P\sqrt{a}/tw)\ (1 - 0.1 \times 2a/w + (2a/w)^2 \ldots)$$

where P = the fracture load

t = thickness of specimen

w = width of specimen

a = 1/2 of notch size

Results

Specimen conditions and results are summarized in Table 1.

TABLE 1

Results of Fracture Tests on Boron-Aluminum Composites

Specimen	Heat Treatment	Thickness t(in.)	Width w(in.)	Notch Length 2a(in.)	Fracture Stress σ(10^3 psi)	Composite Stress Intensity Factor K_{Ic}(10^3 psi$\sqrt{in.}$)	Filament Stress I.F.* K_{If}(10^3 psi$\sqrt{in.}$)
E1	MT6	0.0322	1.065	0.271	80.2	42.0	19.9
E3	MT6	0.0732	1.065	0.259	67.0	33.7	11.0
E5	MT6	0.1010	1.012	0.265	63.9	31.4	10.5
E2	T6	0.0316	1.069	0.260	60.5	30.2	11.2
E4	T6	0.0723	1.090	0.263	60.5	31.0	13.7
						Av.=33.7	13.3

* For the load at which the first filament fracture occurred.

It is noted that the composite stress intensity factors have a normal spread, averaging 33.7 x 10^3 psi$\sqrt{\text{in.}}$ (A careful analysis indicates K_{Ic} = 30,000 ± 4000.) The average K_{If} calculated based on the load at which the first filament fracture occurred is 13.3 x 10^3 psi$\sqrt{\text{in.}}$, about one-third of the former. Incidentally, during the course of the test, filament fracture was detected by recording acoustic (stress-wave) emission monitored by a piezo-electric transducer. The emission began at about one-third the fracture load (well before the load-displacement curve became non-linear) and continued unabated until total fracture.

From Table 1, it is seen that heat treatment had a marked effect on the K_I for thin specimens. The explanation for the effect of heat treatment on K_I is not yet known. It is also seen, from Table 1, for specimen thicknesses greater than or equal to 0.10 in., the stress intensity factor (K_I) is very nearly the critical value of stress intensity factor for unstable crack growth, i.e., the K_{Ic}.

Discussion

For isotropic homogeneous materials, the quantity $(K_{Ic}/\sigma_y)^2$ is the characteristic material dimension to which all limiting specimen dimensions (t,a, etc.) are related, (σ_y is yield strength.) [10]. Earlier results for commercial alloys led to the tentative conclusion that the crack length

and specimen thickness should not be less than 2.5 $(K_{Ic}/\sigma_y)^2$. In this work, the constant notch length, a, and the maximum specimen thickness, t, used were 2.5 x $(K_{Ic}/\sigma_y)^2$ and 1.9 x $(K_{Ic}/\sigma_y)^2$, respectively, where $K_{Ic} \cong 30{,}000$ psi$\sqrt{\text{in.}}$, and $\sigma_y \cong 130{,}000$ psi. Thus, the notch length in the composites was within recommended values (for metals), and as the specimen thickness approached the recommended minimum value for metals, K_I for the composites (MT6 condition) approached its limiting value, K_{Ic}. These results indicate that filament-reinforced metals behave within the guidelines developed for metals. Thus, present results encourage one to believe that conventional concepts of fracture toughness for isotropic materials will be useful in characterizing the toughness of filament-reinforced metals.

It appears appropriate to point out here that some limited test data suggested that the conventional fracture mechanics could apply to fiberglass-reinforced plastics (epoxy), whether unidirectionally reinforced [72], or angle-plied laminates. [39].

Now, what did the fibers do to increase/decrease the fracture toughness of the boron-aluminum composite under consideration? To answer this question, it is necessary first to determine the fracture properties of the 6061-T6 aluminum alloy, and some related ones.

Kaufman and Hunsicker [36] gave the following measured data:

7075-T6: K_{Ic} = 23,000 to 32,800 psi$\sqrt{in.}$

Unit Propagation Energy = 180 in.-lb/in.^{2}

2024-T3: K_{Ic} = Not directly given

U.P.E. = 600 in.-lb/in.^{2}

6061-T6: K_{Ic} = not directly given

U.P.E. = 720 in.-lb/in.^{2}

It was also known that the relationship between the plane-strain stress intensity factor and the unit propagation energy:

$$K_{Ic}^2 = 3000 \text{ U.P.E.} + 400{,}000$$

where U.P.E. is in inch-lb. per square inch, and K_{Ic} in psi$\sqrt{in.}$ Hence, the K_{Ic} for 6061-T6 can be calculated:

$$K_{Ic}^2 = 3{,}000 \times 720 + 400{,}000$$

$$K_{Ic} \cong 50{,}000 \text{ psi}\sqrt{in.}$$

Similarly, the corresponding vlue for 2024-T3 is found to be 47,000 psi$\sqrt{in.}$

It is of interest, thus, to observe that plain aluminum alloys have the following fracture properties:

Type	K_{Ic}, psi$\sqrt{in.}$
6061-T6	50,000
2024-T3	47,000
7075-T6	23,000 to 33,000

Therefore, the present composite has roughly the same

fracture toughness as the 7075-T6 aluminum alloy. The K_{Ic} value can be somewhat higher, though, for higher boron fiber volume ratio. Nevertheless, the composite fracture toughness is not as high as the plain matrix material, the 6061 aluminum alloy. The weakening is most likely due to the fact that this matrix is reinforced (in a negative sense) by the rather brittle boron filaments.

This reduction in K_{Ic} generally does not occur to composites which has rather weak matrix (with low K values) as compared to the fiber. Examples are the fiber reinforced epoxy composites. Epoxy has been found to have relatively low toughness. For example, $K_{Ic} = 1600$ psi$\sqrt{\text{in.}}$ for DOW 332 epoxy cured at room temperature. [17]. This works out to a fracture toughness of about 6.0 in.-lb./in.^{2}. It is really low when compared with 600 in.-lb/in.^{2}, which is a familiar number to most engineers who utilize conventional metallic materials.

4.2.3. The Fracture of Simple Anisotropic Plates

The Test and Background Information

One of the first major efforts made in the area of experimental determination of composite fracture behavior was reported by Wu and Renter [71].

The specimens were fabricated from the unidirectionally

reinforced fiberglass composite known as Scotch-ply. Centrally located notches with various orientations, as shown in Figure 4-6, were introduced. The original crack length varied, intentionally, from specimen to specimen. The thickness of each of these specimens remained a constant, i.e., 1/16 in. Other dimensions of the specimen remained within the conventional practice with isotropic materials, and are comparable to those cited in Sections 4.2.1 and 4.2.2.

The loading conditions are schematically shown in Figure 4-6. These include the tension, the pure shear, the combined tension-shear cases.

Results and Discussion

The specimens were tested in a manner similar to classical tensile test. All were loaded to failure, i.e., unstable fracture. The critical load and crack strength at the state of incipient rapid crack extension were recorded. It was found that the crack propagated in a direction co-linear with the original notch for all types of loading considered. Both the first (the direct opening) mode and the second (the forward shearing) mode were observed.

The stress intensity factors associated with the fracture observed in this series of tests were computed. A careful analysis of the test results indicated that the conventional

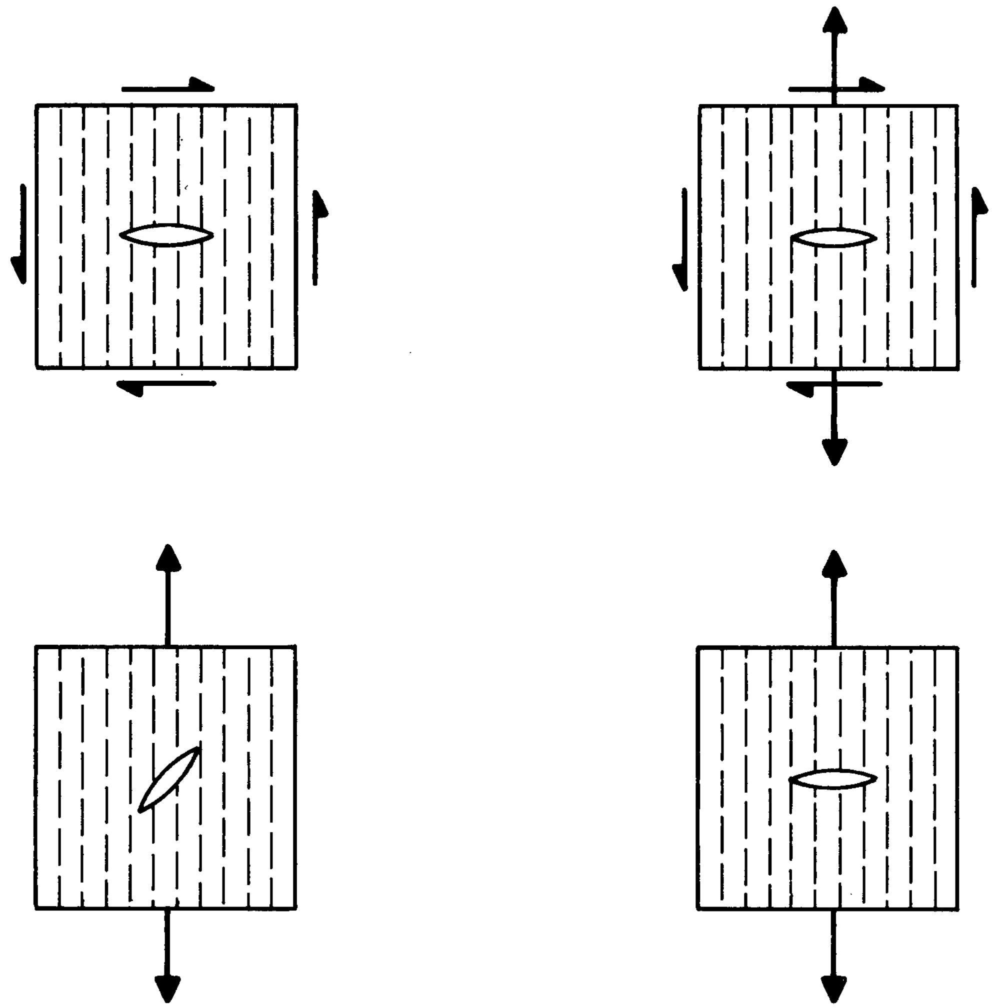


Figure 4-6 Scotch-ply fiberglass - Speciman Geometry and Loads

Criffith-Irwin fracture theory <u>was applicable</u> to the glass composites under study. A lengthy detailed account on the evidence which supported this implication is given in [72]. Naturally, this appears to be a very important conclusion.

4.2.4 Fracture in Laminated Glass-Epoxy Sheet

The Test and Background Information

While the conclusion reached by Wu and Renter [71] was impressive, it was based on results obtained in testing the undirectional glass composite. Its validity for composites other than the scotch-ply was not clear.

Recently, Kendall [39] reported his experiment results on the fracture of angle-plied glass-epoxy sheets with center cracks. His results further substantiated the claim made by Wu and Renter.

As sketched in Figure 4-7, the specimens measured approximately 7 inch x 7 inch. The laminated panels had a varying thickness, depending on the number of plies employed. Some were as thick as 1/8 inch.

The center-notched specimens having various filament orientations (0°, ±60°, etc.) were loaded in tension normal to the notch direction. Recorded was the phenomenological behavior of the specimen, i.e., the initiation and growth of cracks near notch tips, and the final unstable crack failure.

Results and Discussion

Based on the data taken during the lengthy series of

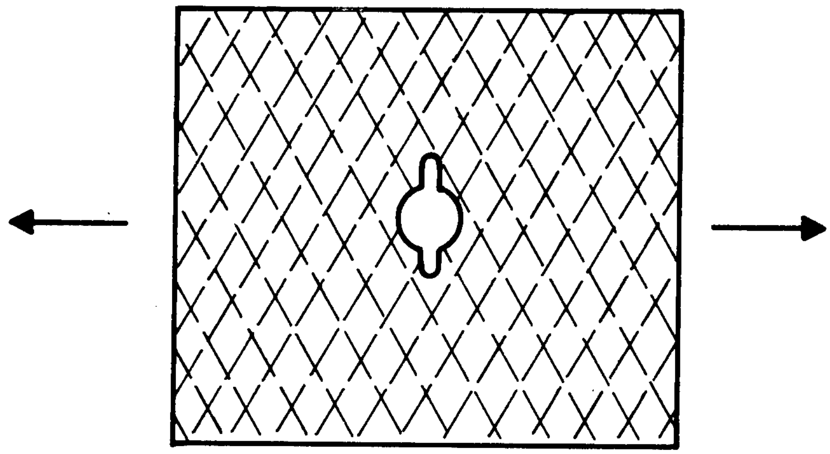


Figure 4-7 The Angle-plied Specimen

tests, stress intensity factors were calculated using theory developed for isotropic materials. Effects of filament orientation, initial notch length, and specimen thickness on the fracture behavior were assessed. It was then concluded that fracture of the angle-plied glass composites could also be predicted using the linear elastic fracture mechanics developed for homogeneous, isotropic materials.

The significance of this work, of course, is apparent. It adds to the small body of experimental evidence which tends to suggest that the classical Griffith-Irwin fracture mechanics is applicable to glass reinforced plastics. Now, the results of [29] on boron/aluminum composite also indicate that the anisotropic composites behave somewhat similar to isotropic material. Thus, it appears that the claim may be a valid one. However, much more experimental

evidence of this nature is needed to substantially verify the validity of this claim. Such has been, at least, the attitude that the fair-minded researcher took with great merits historically.

4.2.5. Fracture of Graphite-Polyimide Composites

The Test and Background Information

The graphite-polyimide composites are currently one of the prime candidates for structural applications in future aerospace vehicles. A test program involving this material was carried out in 1971, and was reported by J. E. Zimmer in 1972 [73].

The composite used in this work was a Modmor-Skybond bidirectional (0°, 90°, 0°) layup. The fiber volume ratio was approximately 60 percent. The fracture toughness tension specimens were designed in accordance with Sprawley and Brown [ASTM 381], with the exception that bonded fiber-glass doublers for loading were used rather than the familiar pin-loading. The finished specimens were 0.025 inch thick, 1.000 and 1.500 inches wide, and 9 inches long with a 3-inch gage length between the doublers. Pre-loading cracks of various length (less than a=w/6) were machined into the specimens by extending the sides of a hole with a diameter of 0.05 inches with a 0.006-inch thick jeweler's saw. That

is the crack tip radius is somewhat undetermined, and is certainly different from that stated in the paper by Sprawley and Brown, where a 0.010-inch jeweler's saw was mentioned. Fatigue cracking of the specimen was not discussed.

All tension tests were conducted at room temperature on an Instron testing machine at a strain rate of 0.006 in./in./min.

Results and Discussion

The fundamental plane strain equation relating fracture stress to crack length for a through-thickness crack in a finite-width tension specimen is

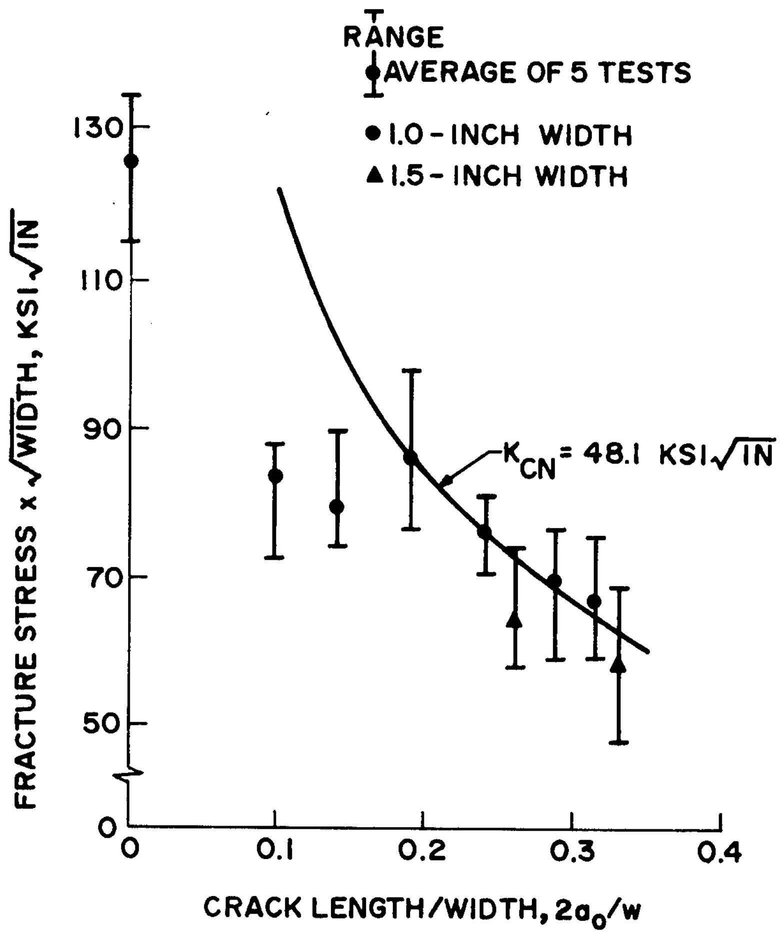
$$K = \sigma \sqrt{w \tan \left(\frac{\pi a}{w}\right)}$$

where K = stress intensity factor

σ = gross fracture stress

a = semi-crack length ($< w/4$)

If the original crack length a_o is used, then the stress intensity factor is K_{CN}. This nominal critical stress intensity factor predicts failure/no failure for a crack which exists in the material prior to the application of load.



Results of the test are shown in above figure. For each set of five specimens, a value of the nominal critical stress intensity factor, K_{CN}, was calculated. The average value of K_{CN} for the 40-specimen test program was 48.1 ksi $\sqrt{in}$. This number was then substituted into the above theoretical equation which was derived for conventional materials. It becomes

$$\sigma\sqrt{w} = \frac{48.1}{\sqrt{\tan \frac{\pi a_o}{w}}}$$

The above equation was plotted in curve form. The preceding

figure shows that a definite relationship exists between the fracture stress and crack length for the present composite and that the relationship agrees with the stress intensity factor equation above. This result is similar to that found by Waddoups, et al. [67]. Their tests showed that failure stress in a graphite-epoxy composite was dependent on a characteristic crack length and a critical stress intensity factor.

Consequently, it can be said that the analyses and procedures for the fracture mechanics of conventional materials can be directly applied to the fracture analyses of this bi-directional graphite-polyimide composite, since the critical stress intensity factor for this composite was constant. Further investigation is, or course, required to determine if this conclusion is valid for other composites with different fiber orientations and constituents.

Incidently, the author had an explanation for the two sets of values which did not follow the curve in the figure presented herein.

4.2.6. Additional Composite Tests of Significance

Generally speaking, reports on fracture failure and other modes of failure are of value in this investigation. In addition to these five reports presented heretofore, a

dozen or so such reports have been received. Those bearing particular significance to this work are briefly discussed below.

Macroscopic Fracture Mechanics of Graphite-Epoxy Laminate

Waddoups and his associates at General Dynamics at Fort Worth conducted a series of static as well as fatigue tests on graphite-epoxy laminates of various stress concentrations. [67]. These included [0°/90°] and [0/±45°] laminates with a range of circular holes and notches. The average static strength of the composite was 83,500 psi for the [0°/90°] specimen which was 1 inch in width and 9 inches in length. Average residual strength after 5 x 10^6 cycles was 77,200 psi.

Test data obtained was analyzed. It was found that the experimental behavior was consistent with the behavior as predicted by the traditional fracture mechanics analysis based on Griffith-Irwin theory.

Fracture in a NARMCO Graphite-Epoxy

Konish and his associates carried out a series of fracture tests with the objective to determine whether the concepts of the conventional linear elastic fracture mechanics might be used to describe behavior of advanced filamentary

Composite laminates. [41,43]. The limited results presented showed a positive finding for the particular laminate in question.

The material used consisted of Morganite II graphite fibers and 5206 resin. General Dynamics Fort Worth Division fabricated the three-point bend specimens. All tests were performed at room temperature in accordance with the ASTM Tentative Method (E399, Part 31 of ASTM Book). The only exception being that the crack front was not sharped under fatigue loading. Instead, the notch was produced by a saw-cut followed by a final lengthening and sharpening using an ultra-sonic cutter.

	K_Q, lb/in$^2\sqrt{in}$ x 10^{-3}			$\bar{K}_Q$ lb/in$^2\sqrt{in}$ x 10^{-3}	G_Q, in-lb /in^2
fiber orientation angle, α	a=0.2 in	a=0.4 in	a=0.6 in		
0°	___[1]	28.8	36.3	$32.6^{+11\%}_{-11\%}$	117.0
90°	1.66	1.46	___[2]	$1.56^{+6.3\%}_{-6.3\%}$	0.943
45°	0.690[3]	2.22	2.39	$2.30^{+3.9\%}_{-3.8\%}$	___[4]
$(\pm 45°)_s$	18.5	18.5	16.3	$17.7^{+4.8\%}_{-9.4\%}$	45.0
$(0°/\pm 45°/90°)_s$	23.5	21.7	20.5	$21.9^{+7.3\%}_{-8.6\%}$	55.1

Note:

1. Specimen was crushed before crack propagation occurred.
2. Instrumentation failure.
3. This value omitted when calculating $\bar{K}_Q$.

4. No G_Q available because the crack propagated in a mixed mode, which could not be directly uncoupled.

5. K_Q is the critical stress intensity.

6. $\bar{K}_Q$ is the average value of several K_Q values for a laminate.

Shown above are experimental results obtained from the test program. It is clear that fiber orientation played a major role in fracture resistance.

Tensile Fracture Mechanisms in Carbon-Epoxy

Mullin and Mazzio conducted a series of tests for the purpose of identifying failure mechanisms for some of the more widely used carbon fibers in epoxy under tensile loading. [50]. The approach was first to encapsulate single fibers and bundles of fibers in the epoxy matrix and observe their failure mechanisms. The observations would then be utilized to analyze the failure process in more heavily reinforced composites.

The behavior of three kinds of carbon fibers was studied: high strength, high modulus and low modulus fibers. The resin matrix selected was DEN 438, an epoxy novolac capable of being modified to change its sensitivity.

All three types of carbon fibers were found to have failed by cleavage of the resin at the first fiber fracture in the unmodified resin. The fracture initiated at the

fibers and propagated through the resin. There was no evidence of debonding. In contrast to this behavior a number of cleavage cracks and debonded regions were observed in the modified resin which had greatly improved toughness (nearly 3 times). Moreover, additional observations were made of gross failure modes in composites involving various carbon fibers. The ASTM test method D-638-38 was followed.

Fracture Mechanisms in Boron-Epoxy

In 1968 Mullin, Berry and Gatti published their experimental findings on the fundamental fracture mechanisms in the boron-epoxy composite. [49]. Some interesting results were obtained. Observed failure modes were somewhat similar to other filamentary composites. Naturally, it is premature to make a general statement regarding boron composites, as very little test data can be found in open literature at present. The fact that boron fibers usually are much larger in diameter than graphite fibers is sufficient reason which warrants further investigation.

Fracture Toughness of Boron-Aluminum Composites

Among the first formally published test programs involving filamentary composites was one conducted by Adsit and Witzell in 1969. [1]

Diffusion bonded boron-aluminum composite specimens, both notched and un-notched, were manufactured. The matrix is 6061 aluminum, and the filaments were standard 4-mil boron with no coating. Three different types of specimens were used: (1) unidirectional specimen with 25 percent fiber volume, (2) unidirection type with 50 percent fiber volume, and (3) 0°/90° cross-ply type with 45 percent fiber volume. Two different material thicknesses were investigated: 4 ply (nominally 0.020-inch thick) and 16 ply (0.080-inch thick).

The specimens were tested in an Instron testing machine at a head rate of 0.010 inch per minute. The apparent plane-strain fracture toughness values from the loads obtained by the 5 percent secant method, as well as the loads at pop-in, were obtained by using the following equation:

$$K_{Xc}^2 = \left(\frac{P}{B}\right)^2 \frac{1}{w}\left[7.59\left(\frac{a}{w}\right)^2 - 32\left(\frac{a}{w}\right)^2 + 117\left(\frac{a}{w}\right)^3\right] \frac{1}{1-\nu^2}$$

where P = load

w = specimen thickness

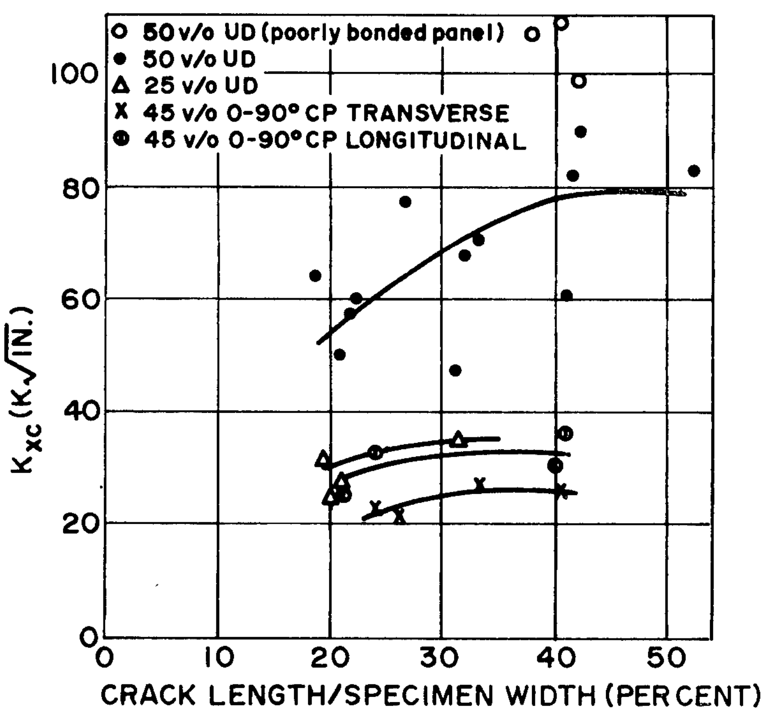
B = specimen width

a = notch and crack length

ν = Poisson's ratio

The results on fracture toughness of the series investi-

gated are shown below. The scattering nature of test data is clearly visible. The 25 percent unidirectional material and the 45 percent cross-ply material are seen to have a fracture toughness that is in the range typically quoted for aluminum alloys when the upper load pop-in value is used. [35]. The 50 percent unidirectional specimens seem to possess higher fracture toughness values. Additionally, they seem to indicate that fracture toughness varies with the ratio of crack length to specimen width. Observations made here, however, require further substantiation as the computation of K_{Xc} was made with the assumption of a great many things.



Fracture Modes in Aluminum-Matrix Composites

Ruzauskas and Hay published results of their recent investigation into the mechanisms of fracture in two composites and their dependence on loading conditions in various states of static loading. One of the composite systems was made of continuous stainless steel wire (0.009 inch diameter) and 2024 aluminum matrix. The wires which had a minimum strength of 475,000 psi were introduced into the matrix through diffusion bonding, resulting in composite plates 0.20 to 0.25 inch thick. The other composite material was a tungsten-aluminum system. [55]. Tungsten wire 0.005-inch diameter with a strength of 415,000 psi was used. Both are unidirectionally reinforced. Test specimens were fabricated according to applicable ASTM specifications. These specimens were tested in five stress states including longitudinal tension and compression, transverse tension and compression in the plane of the plate, and a through-thickness compression.

Many scanning electron fractographs and optical micrographs characterizing failure modes in various testing modes in both systems were obtained. Yielding, work hardening and fracture were observed in the longitudinal tension tests. In transverse tension specimens, fracture preceded yielding. In the SS-Al system a typical dimpled rupture of the matrix was observed between the filaments while fracture propagated

around the filaments by delamination at the fiber-matrix interface. The W-Al system exhibited similar matrix fracture features, but fracture propagated through the filaments by the longitudinal filament splitting.

As tested in longitudinal compression, both systems exhibited yielding prior to failure by the shear mode of in-phase buckling. The SS-Al system showed considerable delaminations while this was absent in the W-Al system.

In transverse compression tests yielding was observed in both composites. The SS-Al specimens failed by shear on the planes of maximum shear stress, and fracture propagated around the filaments by filament-matrix delamination. In constrast, the W-Al specimens failed by matrix-matrix delaminations.

Deformation of SS-Al specimens in the through-thickness compression test was nearly identical to that in transverse compression. The W-Al specimens failed on the planes of maximum shear stress with fracture proceeding through the fibers by longitudinal splitting.

These observations clearly illustrate the multiplicity of failure mode which can exist in a single composite system and their dependence on the mode of testing. It is only logical to expect further complexities in failure modes when

angle-plied specimens are introduced, not to mention varying types of composites.

5. FRACTURE OF COMPOSITE MATERIALS

The filamentary composite material under investigation has its unique mechanical behavior. In general, its high strength to weight ratio makes it a potentially very useful material for flight vehicles for which weight is a very important consideration. However, the fracture behavior of such composites has not been very well understood yet. It goes without saying that caution must be exercised by the users at this stage of development.

In the sections which immediately follow, an attempt will be made to discuss some important aspects of the fracture of fiber-reinforced composite materials.

5.1. Significance of Fracture

Fracture has been observed in engineered products made of either conventional (homogeneous, isotropic) materials or filamentary composites. Some of the fracture failures in recent years have had profound effects upon not only some products but also the business success of some giant industrial concerns.

About five years ago, a famous company in Europe which had a long history of success decided to utilize the seemingly attractive carbon-fibre composite for a new high-performance jet engine. Some technical information about the project

can be found in a book by R. M. Gill. [25]. It was published in 1972. The book, however, appeared to be written in perhaps 1971. Now, it was unfortunate that this development effort turned out to be an extremely costly exercise for the company. The failure brought about the bankruptcy of the company. It also made another industrial giant, an American airframe manufacturer which designed a new jet airliner around the new engine, come very close to the point of collapse.

Although details of the failure are lacking, it is a considered opinion of many in the engineering field that poor fracture toughness of the graphite composite material selected by the designer was responsible for the failure.

As discussed in Section 2, the mechanical behavior of filamentary composites are unique in many aspects. Of significant interest to a designer are the high strength of the material, and the fact that material imperfections do exist. Consequently, fracture failures do represent a distinctive area of concern for those who plan to utilize the composites. Unfortunately, the subject matter is relatively new and rather complex. There have been only dozens of meaningful investigations on fracture of composites over the years. The limited volume of experimental data gives a good indication that this branch of material engineering is still in its infancy and that much work remains to be done.

5.2. Nature of Fracture in Composites

It is well known that members made of filamentary composites usually contain flaws. Consequently, the possibility, or probability, of a tensile fracture in such members does exist.

Past engineering experience with isotropic materials indicates that flaws exist in the form of cuts, gouges, nicks, air bubbles, foreign (weak) particles, or microcracks. When loaded in such a manner that tensile stresses developed nearby, these flaws become the source of initial microcracking. These microcracks grow with increasing loads. As the cracks grow into macrocracks, they become more visible and potentially dangerous. The crack tips were found to have plastic zones of limited size. Therefore, the propagation of the cracks is irreversible, i.e., the cracks will not self-seal. This type of stable crack growth continues until the rapid, or unstable, crack takes place, when the member completely fails. Thus, the fracture in homogeneous isotropic material involves three phases: (1) the initiation of cracks, (2) the stable crack growth, and (3) the final unstable crack failure.

As discussed earlier in Section 3, a filamentary composite material is non-isotropic and heterogeneous. As such, it appears that the fracture process in composites may not

be the same as that of conventional materials. The truth, however, seems to be just the opposite, based on the limited experimental evidence gathered to date. More discussions on this will follow.

Lastly, as the composites are anisotropic, it is expected that, in a simple laminate, there are at least six fracture toughness (G in in.-lb./in.^{2}) values. These include three G's (G_I, G_{II} and G_{III} for the I, II, and III fracture modes, respectively) for the longitudinal direction, and a still different set of 3 G's for the transverse direction. In addition, there remains the question of whether the 3 G's associated with the lateral (across the thickness) direction should be pursued. Consequently, it can be seen that it will not be inexpensive to fully understand the fracture process in composite materials. Very frequently, however, composites are utilized to resist tension at an extremely high stress level. In light of the existence of flaws, it is mandatory that the question of fracture toughness be considered in such a design, even if it is somewhat costly. This task becomes all the more important when a state of high tensile stress exists.

5.2.1. Stress Concentrations and Material Imperfections

Stress concentration in a structural member refers to the complex pattern of stresses (normal as well as shearing)

which normally exists in the vicinity of a hole, a groove, or a cut or crack, or any sharp change in geometry of the member. It is a stress buildup. In a structural member made of conventional material, the stress concentration factor is usually somewhere between 1 and 3. Is this also true of members made of filamentary composites? The answer is wanting, as little is known of stress concentration in composites.

As pointed out in Section 2, the stress concentration factor is directly related to the stress intensity factor. The latter governs fracture toughness of a material. It is clear, then, that stress concentration is also a governing factor in the fracture behavior of a material, be it a conventional type or the composite in question. The need to further understand stress concentration in composites becomes evident here.

The presence of a hole, or a cut, in a composite member introduces an additional complicating factor: the presence of discontinuous fibers. In a composite laminate, for example, the introduction of discontinuous fibers changes the very nature of the fiber-matrix construction. Naturally, this will result in a corresponding change in fracture toughness.

The presence of material imperfections in a composite

is without question. These exist in the form of cuts, nicks, voids, poor fibers, poor bonding between the fiber and the matrix, uneven distribution of fibers, etc. The net result is a reduced fracture toughness. While it is possible, and desirable, to reduce the amount of imperfection, it is not possible to eliminate such imperfections all together. Therefore, fracture is always a possibility in a tensile-stressed member. [67].

5.2.2. Governing Factors in Fracture

For a material, there exists a fracture toughness index. This is a well established fact. Thus, the matrix of a composite material possesses a toughness index. So does the fiber. For example, the Dow 332 epoxy matrix is said to have a stress intensity factor K_{Ic} of 1600 psi $\sqrt{in.}$ [17]. The 7075-T6 aluminum alloy has a K_{Ic} of 23,000 to 33,000 psi $\sqrt{in.}$ [36], whereas that of 6061-T6 aluminum alloy is approximately 50,000 psi $\sqrt{in.}$

Now, the composite material is manufactured from the constituent matrix and fibers. It is logically true that the fracture toughness of the composite depends on the fracture behavior of its constituent materials. This statement has also been found true in numerous composites. Moreover, the actual resistance to fracture in composites does vary from one composite system to another. Indeed, the composite

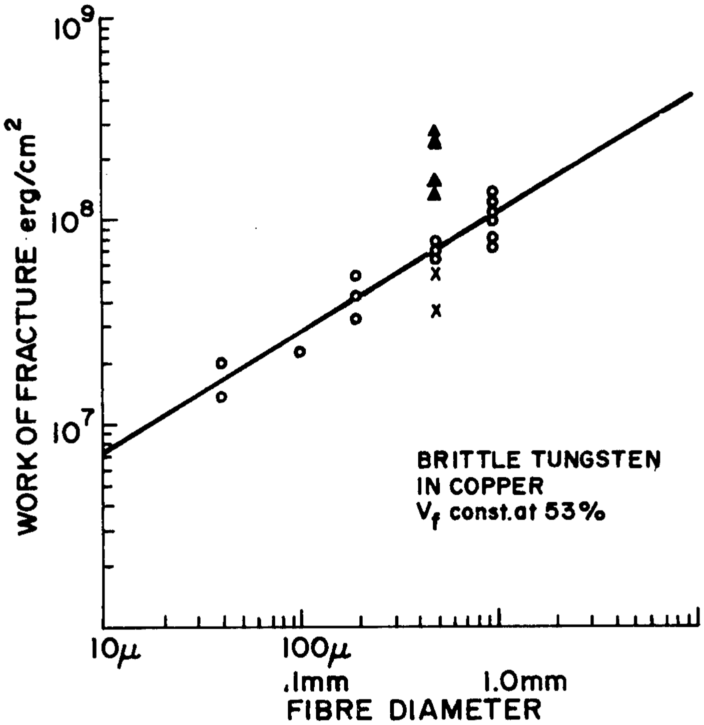
fracture toughness may also be dependent on other important factors. Some of them are given below.

Fiber Diameter

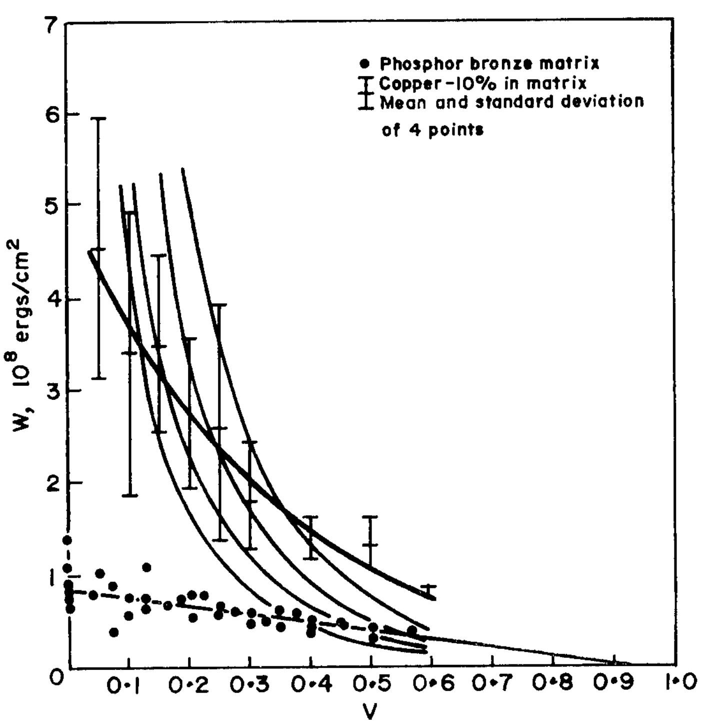
A. Kelly reported in late 1971 that work of fracture of a composite is better with large (greater than 25xμm) diameter fibers whereas thin fibers reduce cracking or yield of the matrix and improve fatigue and creep resistance of the composite. In a different paper, R. E. Cooper reported in 1970 that composite toughness increased with increased fiber diameter. Therefore, it is seen here that fiber diameter is becoming recognized as a parameter as important to the composite technologist as is grain size to the metallurgist. Furthermore, these statements also point to the fact that composite toughness and tensile strength are conflicting properties. A similar behavior for conventional materials also exists, as discussed in Section 2. Shown is result of a series of three-point bending tests performed on specimens made of tungsten fiber-copper composite material. The matrix was vacuum-cast. [14]. Clearly, the work of fracture increases rapidly with increasing fiber diameter. (The figure is on the next page).

Fiber Volume Ratio

From Cooper's [15] test results on specimens made

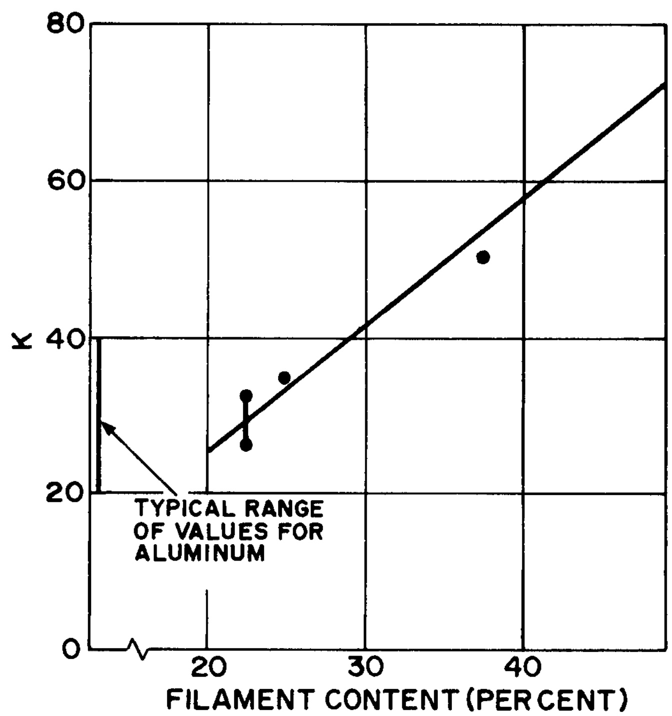


of (brittle) tungsten fibers and (ductile) copper matrix, it was clear that fiber-volume ratio is an extremely important factor in fracture toughness of a composite. The following curves



serve to indicate the significant effect of the volume ratio on the work to fracture.

A somewhat earlier (1969) report by Adsit and Witzell, however, gave a somewhat different account on this, based on their findings with the boron-aluminum composite [1]. They indicated that apparant fracture toughness varies with the filament content in the manner shown below.



Fiber Length

In a series of experiments, G. A. Cooper [13] found that the greatest toughness for a given strength is always found in composites reinforced with discontinuous fibers.

Phosphor-bronze was used for the reinforcing fiber since it is very ductile. The brittle HY951 resin was the matrix. In the tests, fiber fractures occurring near, but not at, the plane of matrix failure in the composite, led to fiber pull-out during fracture. Energy absorbed in this process was thought to have contributed directly to the work of fracture and hence to the toughness of the composite.

It should be recalled here that a single fiber, in a unidirectionally reinforced composite, considered in isolation, is expected to pull out if one end is within a distance $\ell_c/2$ of the fracture plane of the matrix, where ℓ_c is the "critical length". If not, the fiber will be broken. Here, the discontinuous fiber was used in the composite.

A composite reinforced with parallel, continuous fibers which have no weak points is expected to fail on a plane normal to the fiber axis, and there will be no fiber pull-out. Thus, the resulting work-to-fracture is lower than that of composites with discontinuous fibers of the same strength.

Matrix Strength & Ductility

When a unidirectionally reinforced composite is loaded parallel to the fiber axis, fracture initiates either by

a) matrix tensile fracture

b) matrix shear fracture

c) interface shear fracture (debonding)

d) fiber fracture

depending on which of these processes occurs first, i.e., at a lower value of composite stress (applied strain). The first two fracture initiation modes are directly related to the strength of the matrix. For example, the 6061-T6 aluminum matrix is known to have much higher strenth as well as ductility than the epoxy resin matrix. That matrix properties have significant influence on composite fracture is clear.

A comparative study of tensile fracture mechanism can be found in Reference [50].

Composite Bonding

As pointed out in above paragraph, composite fracture can initiate as a result of debonding. Bonding between the fiber and the matrix usually depends on the physical properties of composite constituents involved. Nevertheless, the actual interface shear distribution is not well understood at present. Same is true of the shear stress concentration at the end of the fiber.

Excellent bonding can be achieved with metal matrices. In general, matrices offer moderately low (good) interface bonding.

Orientation of Fibers

It is well known that the strength characteristics of composites depends heavily on the orientation of fibers. Same is true of fracture toughness of composites. Many test results substantiated this observation. Some authors consider the orientation of fibers the single most important factor in fracture toughness. [42], [67].

Environmental Conditions

That environmental conditions have significant influence on the fracture toughness of many plastic type materials is a common knowledge. It is also true of composites. Particularly, temperature is thought to be a most important factor. Moisture and salt are also influential in some glass epoxy composites.

Other Significant Factors

As pointed out in Section 2, the thickness of test specimen has its influence on fracture toughness of the isotropic homogeneous specimen. It can be logically expected that thickness is a factor governing fracture toughness of filamentary composites.

The relative ductility or brittleness between the fiber and the matrix is also thought as a contributing factor as

far as composite toughness is concerned. Much more work needs to be done before this factor is well understood.

Residual stresses are thought (not without evidence) to exist in most composite materials. Little is known of the state of such stress nor its significance.

5.2.3. Fracture Initiation in Composites

There are two principal types of filamentary composites: the composite with "continuous" fibers and the one with "discontinuous" fibers. They exhibit different behavior, even in the absence of flaws, when subjected to tensile loads.

In a unidirectionally reinforced continuous composite material, the filaments transmit load directly. In the discontinuous composites, the axial tensile load is transmitted to the fibers through shear stresses developed at the matrix-fiber interface. Consequently, the tensile resistance in these two cases is governed by different factors.

For composites with unidirectional continuous fibers the nominal stress carried by the composite is:

$$\sigma_c = [E_f V_f + E_m(1-V_f)]\varepsilon \qquad (5.1)$$

where ε = strain, E_f = fiber modulus of elasticity, and E_m = matrix modulus of elasticity. The above equation essentially governs the initiation of fracture in a composite with continuous fibers.

Now, the interface stress developed in a composite system involving discontinuous fibers causes a tensile stress in the fiber. This stress is a maximum at or close to the end of the fiber, and then decreases to zero at the center of the fiber. [14]. Approximately, the fiber must have a length

$$\ell_c = \frac{d}{2}\frac{\sigma_f}{\tau} \tag{5.2}$$

before it can be fractured. Here d = the diameter of fiber, σ_f = critical (max.) tensile stress of fiber, and τ = the maximum shear stress that can be supported in either the matrix or the interface, which ever is smaller. The above equation can be rewritten as

$$\frac{\ell_c}{d} = \frac{\sigma_f}{2\tau} \tag{5.3}$$

where ℓ_c/d is defined as the critical aspect ratio. When $\ell = \ell_c$, the average stress, $\bar{\sigma}_f$, carried by the fiber is equal to $1/2\sigma_f$. When the fiber length is much greater than ℓ_c, the average stress in the fiber approaches σ_f. The tensile stress carried by the composite is then (σ_m is the stress in matrix):

$$\sigma_c = \bar{\sigma}_f V_f + \sigma_m(1-V_f) \tag{5.4}$$

At this load, the composite material faces the danger of a tensile failure.

The above discussion is concerned with the tensile fracture strength of the "ideal" filamentary composite material. This material is assumed to be free of flaws. In reality, however, it is almost unavoidable to have flaws in composites. These flaws tend to decrease fracture resistance of the material.

Experimental results presented in previous sections tend to support the thesis that the fracture behavior of composites with flaws is governed by the linear elastic fracture mechanics. However, this "conclusion" is tentative. It is true that it applies to the cases reviewed. But there is no way of determining if it is also applicable to composites in general. Obviously, more verification tests must be conducted before the thesis can become a useful criterion governing the fracture behavior of composites.

5.2.4. Crack Growth and Fracture Failure

Naturally, cracks (initiated because of the presence of flaws) grow as the tensile stress builds up. There is little information on this subject matter. From the dis-

cussion presented in Section 4 it appears true that the cracks grow stably and progressively. Cracking, splitting and branching can happen, depending on the properties of the composite material under consideration.

Now, as discussed in last section, fracture can take place in the form of fiber fracture. (Assuming continuous fibers or discontinuous fibers with $\ell > \ell_c$.) Several events can occur, depending on the properties of the matrix, the fiber, and the interface.

Case 1. <u>All filaments have the same strength</u>

If all filaments have the same σ_f, then the failure of all the filaments at the same stress level [Equations 5.1 or 5.4] immediately leads to the unstable composite failure. This type of behavior was observed when metal fiber-metal matrix composites were broken in unaxial tension. Examples are Mo-Cu, W-Cu systems, [34, 37].

Case 2. <u>Filaments have a distribution of strengths and matrix is brittle or semi-brittle</u>

This type of behavior was found to have occurred in composites with non-metallic filaments imbedded in certain epoxy resins. [33]. Section 4.2 offers good examples. It is known that many resins fracture in a relatively ductile manner at low strain rates and moderate temperatures, i.e., large amounts of necking preceed final fracture, and in a

brittle manner at high strain rate and low temperature. Therefore, the strain rate and temperature appear to be important. Now, the sudden fracture of a filament imbedded in a matrix of this type will cause the matrix adjacent to the fiber break to be rapidly loaded. The sudden release of stored energy in the filament will be available to allow the fiber crack to propagate into the matrix. The crack continues to grow, as loading increases, until the crack reaches the next filament, causing it to fracture. This process continues until the specimen breaks completely.

Now, a few words about the strength of the filaments. In the absence of flaws or man-made notches, the weakest filament is likely to fail first when the composite specimen is subjected to the same stresses throughout. Examples are so numerous that it is only necessary to re-emphasize the desirability for high standards of quality control.

Case 3. <u>Filaments have a distribution of strengths and matrix is relatively tough</u>

This kind of behavior occurs for non-mettallic filaments in tough resin or metal matrices. Some of these were presented in Section 4., i.e., the boron-aluminum. The glass-epoxy system is another. It has been found that crack through such matrix was retarded or prevented, largely because of debonding between the fibers and the matrix. Thus, the resulting composite fracture strength is, more or less,

similar to that of Case 1.

The work of tungsten fiber/copper matrix by Cooper and Kelly affords a good example. [63]. Fibers ahead of the advancing crack front were found broken and the remaining bridge of matrix then necked down and fractured in a typical ductile manner. It so happened because the filaments were very brittle, and the matrix was relatively ductile. Consequently, the composite toughness resulted from the work done in drawing down the matrix material. The introduction of these "reinforcing" fibers which were brittle, did not enhance the fracture toughness of the composite.

For such a system, it was found that [63] the work per unit length of crack growth, G_{Ic}, is

$$G_{Ic} \cong \frac{(1-V_f)^2}{V_f} d \, \sigma_m \, \varepsilon_m \tag{5.5}$$

where σ_m = the ultimate strength of the matrix

ε_m = the uniform elongation of the matrix

d = the diameter of fiber

V_f = volume ratio of the fiber

For the W-Cu composite system, $\sigma_m \varepsilon_m$ = 10,000 in. - lb/in^2. Thus, its fracture toughness, G_{Ic}, is also dependent on V_f and d. The need for a larger fiber diameter is thus clear. Above all, it can be seen, from Eq. (5.5), that the composite

fracture toughness is closely tied to that of the matrix. In general, the matrix toughness is relatively low when the matrix has a high yield strength and a high volume fraction of voids or other flaws. Consequently, while unnotched composite strength can be increased if high yield-strength matrices are utilized, this gain can be more than offset by the decrease in G_{Ic} as a result of flaws.

5.3. Fracture Toughness Prediction

In Section 5.2., the nature of composite fracture was discussed from test and micromechanics point of view. Some paragraphs were devoted to the analysis of toughness as a function of several factors, such as fiber diameter, fiber volume ratio, etc. These items are certainly of interest to materials engineers or designers who are faced with the task of assessing the behavior of a selected composite material. Alternately, the designer can use his knowledge on composite fracture behavior in designing the composite system itself to meet a performance requirement.

Naturally, any discussion on fracture of structural members made of composite materials remains incomplete without mentioning fracture toughness prediction. The following five subsections, 5.3.1. through 5.3.5., are devoted to this very subject. In these discussions, it is assumed that the structural composites are loaded, at room temperature, by

static loads only. Furthermore, the discussion is limited to the first fracture mode - the direct opening mode. Little information is available on the fracture of composites in the remaining modes.

5.3.1. Some Experimental - Analytical Work

Several published reports related to experimental investigation of the fracture of composites were reviewed in Section 4 of this work. The authors essentially followed test procedures recommended by ASTM publications. Some attempts were also made to analyze test results obtained within the context of traditional LEFM. Undoubtedly, some of these undertakings can serve as a guide, to the designer, in predicting fracture toughness of comparable composites.

Interested designers should also refer to several other publications which offer different opinions and methods regarding fracture of filamentary composites. These include several papers by M. R. Piggott [54], G. A. Cooper [13], G. P. Anderson [2], and M. E. Waddomps, et al [67].

5.3.2. Applicability of Griffith-Irwin Theory

It has been well established that the linear elastic fracture mechanics based on the Griffith-Irwin theory is applicable to conventional engineering materials. The

matured state of development of LEFM, as well as importance, is evidenced by the adoption of structural fracture criteria in the design and development of the new strategic bomber, the B-1. [12,3]. Thus, knowledge on fracture acquired over the past several decades is being incorporated into the complete airframe design for the first time in history. It is felt that LEFM cannot but help insure aircraft structural integrity. Whether it will prevent undesired fracture failures which had occurred on many occassions (to a number of airplanes) in the past, and yet at a reasonably acceptable weight penalties remains to be seen.

Now, the questions of whether the Griffith-Irwin theory is applicable to the fracture of filamentary composites becomes very important, in light of the current desire in effectively utilizing composites for high performance flight vehicles.

During the course of this work, an effort has been made to analyze all available results of numerous experimental investigations reported in open literature from world over. It is believed that the Griffith-Irwin approach is applicable to the composites for the types reported in this work. In fact, several authors have indicated so, largely based on results of individual fracture test programs which were somewhat limited in terms of numbers as well as material parameters. Nevertheless, a trend has emerged to indicate

the applicability. Furthermore, it is thought, in the opinion of this writer, that LEFM has the potential to closely describe the fracture behavior of most filamentary composites known today. However, much more testing work remains to be performed before such hopeful predictions can be verified.

M. E. Waddoups and his colleagues [67] investigated the static and fatique fracture behavior of flawed graphite-epoxy laminates with various fiber orientations: [0/90°] and [0/±45°]. The flaw was a circular hole. They were able to predict the observed fracture behavior by using the LEFM, recognizing that fracture toughness (the G-value) varied with laminate configuration.

H. J. Konish and his colleagues [41,42,43] performed a series of tests on flawed fracture specimens of graphite-epoxy. Fiber orientations were again [0/90°] and [±45°]. Experimental results, as well as subsequent analyses in three papers by the same three authors, indicated that LEFM was applicable to the particular composite specimen tested.

From the above, it is clear that the limited amount of experimental data on fracture does indicate that LEFM can be extended to cover the fracture behavior of composites. A discussion of the fracture toughness for composites, based on LEFM, will be given in the next section.

5.3.3. Fracture Toughness for Composites

Most filamentary composites (laminates) in use today are not isotropic, but are arranged symmetrically about the X-Y orthogonal coordinate axes. Most are mid-plane symmetric. Since, in problems of fracture, stresses which lie outside the plane of the sheet can be neglected, such composites can be considered orthotropic. This class of composite materials includes laminates containing unidirectional fibers, and those with fibers which are parallel to any two directions in the laminate, not necessarily at right angles.

In a paper published in 1965, Paris, Sih, and Irwin showed that stress intensity factors for an orthotropic material are identically the same as for an isotropic material. That is to say, for composites, the following relations hold: [51].

$$K_{Ic} = \sigma_f\sqrt{a} \quad \text{for plane stress} \tag{5-6}$$

$$K_{Ic} = \sigma_f\sqrt{\pi a} \quad \text{for plane strain} \tag{5-7}$$

Consequently, the fracture stress, σ_f, can be determined for a selected flaw size in a composite material whose K_{Ic} is known. For the designer, the task becomes one of making a proper decision regarding the maximum acceptable flaw size. In doing so, he should consider not only the quality of the

material, but also the flaw size which can be well detected in a non-destructive test of his product.

Naturally, the designer is also interested in the stress intensity factor, K_{Ic}. Normally, he may determine it by experimental methods for a selected composite material system. Alternately, he can design a material taylored to his needs by considering relevant aspects of fracture behavior discussed in Sections 4.2 and 5.2. For this purpose, the following factors, among others, are considered:

Fiber Orientation
Fiber Diameter
Fiber Types (strength and ductility)
Fiber Volume Ratio
Fiber Length
Matrix types (strength and ductility)
Composite Bonding
Manufacturing Methods
Workmanship and Workability
Residual Stresses
Environmental Conditions
Nature of Loading in Service
Size of Part in Question

The importance of each of the factors listed above, as far as the fracture toughness of a composite laminate part is concerned, varies from case to case, largely depending on

the nature of the application. However, experimental evidence (and logic) suggests that the variation in fracture strength is, to a great extent, a function of the fiber orientation, once the material system is characterized. Logically, fracture toughness values for different orientations must be properly determined.

Above all, the designer usually has as his goal to devise a composite system with optimized performance - strength, fatique, fracture, thermal, etc. He certainly should strive to use a good material with a reasonably high value of G_{Ic} (or K_{Ic}). For metallic materials, a "magic" G_{Ic} range of 100 to 600 in-lb/sq in is usually encountered. The corresponding range of K_{Ic} is approximately 20,000 to 200,000 psi$\sqrt{in}$. Will composites possess comparable fracture toughness? Some discussions on this will be given in Section 5.5.

5.3.4. Test and Analytical Predictions

It has been well established that G_{Ic}, or K_{Ic}, is a material property, just as Poisson's ratio, the yield stress and the ultimate strength are. The latter ones are ordinarily determined by mechanical tests in a laboratory, by either the user or the material manufacturer. Therefore, it is quite logical that fracture toughness values, for a filamentary composite material being considered in a part design, should

be obtained by reliable experimental method.

Alternately, the designer can resort to analytical methods in predicting the G_{Ic} (or K_{Ic}) by using a few formulas and the basic material constants which characterize the behavior of individual constituent materials involved. As an example, he may use the following equations:

$$G_{Ic} = d\ (1-V_f)^2\ \sigma_m \varepsilon_m / V_f$$

(See page 5-18)

for unidirectional composites with non-metallic filaments in tough resin or metal matrices, such as the boron-aluminum or the glass-epoxy. To obtain the fracture toughness, the basic data needed are fiber diameter, volume fraction, the ultimate strength of the matrix, and the uniform elongation of the matrix.

The above formula represents the type of equations which have been obtained from limited experimental work reported in open literature [52, 56, 63]. Nevertheless, the practical usefulness of these formulas must be subjected to close examination, as they were based on limited results of special tests (often pilot in nature).

In general, the scarcity of experimental results is also having its parallel in the analytical treatment of subject matter. The obvious reason is that the latter usually is

heavily guided by the former. The rather limited number of analytical undertakings, however, has helped the advancement of the state-of-the-art. Nevertheless, it will be a few years from now before the science of fracture of composites will grow to its maturity, when the designers can be guided solely by theoretical calculations. At present, there exists little engineering data on fracture toughness of composites. Hence, the designer is forced to resort to development tests which are costly but necessary.

Now, a few words about analytical expressions for orthotropic composites are in order. Consider the directions of orthotropy as the first and second directions of tensile stress and strain, and consider a crack lying along the first direction, being stressed in such a way that only K_{Ic} is non-zero, Paris and his colleagues [51] found that

$$G_{Ic} = K_{Ic}^2 \left(\frac{c_{11}c_{22}}{2}\right)^{1/2} \left[\left(\frac{c_{22}}{c_{11}}\right)^{1/2} + \frac{(2c_{12} + cc_{66})}{2c_{11}}\right]^{1/2} \pi \qquad (5\text{-}8)$$

where the c_{ij}'s (i and j are integers from 1 to 6) are elastic constants in a generalized Hooke's law of the form

$$\varepsilon_i = \sum_{j=1}^{6} c_{ij} \sigma_j \text{ (for } i = 1 \text{ to } 6\text{)} \qquad (5\text{-}9)$$

$$c_{ij} = c_{ji}$$

where ε_i is the i-th component of strain, and σ_j is the j-th component of stress. The elastic constants are related to

the mechanical moduli by following equations

$$
\begin{aligned}
c_{11} &= \frac{1}{E_x} \\
c_{22} &= \frac{1}{E_y} \\
c_{12} &= \frac{-\nu_{yx}}{E_x} = -\frac{\nu_{xy}}{E_y} \\
c_{66} &= \frac{1}{G_{xy}}
\end{aligned}
\qquad (5\text{-}10)
$$

where the subscripts x and y denote the first and second directions, respectively. E_x and E_y are the familiar Young's moduli for the directions noted, and ν_{xy} is the coefficient relating the decrease in extension in the x direction for extension in the y direction, and G_{xy} is the familiar shear modulus. Such relations have been well defined in books on the theory of elasticity of an anisotropic elastic body. [45].

Substitution of Equation (5-10) into (5-8) leads to:

$$G_{Ic} = \pi \frac{1}{E_y} K_{Ic}^2 \left[\left(\frac{E_y}{4E_x}\right)^{1/2} - \frac{\nu_{xy}}{2} + \frac{E_y}{4G_{xy}} \right]^{1/2} \qquad (5\text{-}11)$$

The above equation is analogous to the equations presented in Section 2, which were for isotropic materials. (Eqs. 2-10 and 2-12). It establishes a definite relationship between G_{Ic} and K_{Ic}, when the composite material constants (E_x, E_y, G_{xy}, and ν_{xy}) are known. Thus, G_{Ic} for a composite material, for a given fiber orientation, can be calculated, when the

corresponding K_{Ic} is known. The latter, of course, can be experimentally determined as discussed in Sections 3 and 4.

5.3.5. Thoughts of Non-LEFM

Up to this point, it has been stated many times that basic concepts and techniques of LEFM are applicable to filamentary composite material. It is the concensus of opinion voiced by a number of authors previously indicated. Almost all of them arrived at the conclusions based on limited test results. Evidence of complete vigorous treatment of the subject question has not been found by this writer.

As is well known, the LEFM was developed for the conventional isotropic homogeneous materials. Now, the filamentary composites are heterogeneous. How does one explain away effects of such heterogeneity, if any, upon the fracture toughness? To date there exists no answer to this question. As a matter of fact, some authors [74] did raise such a question.

Furthermore, the fiber debonding in composites have been observed. This failure mode is not expected in a conventional material. Injured or broken fibers tend to introduce further complications as far as theoretical clarification is concerned. Likewise, the fact that matrix cracking may precede fracture of the composite system also raises some

doubts about the general applicability of the LEFM to composites.

E. R. Frye and R. M. Rayner published the results of their work on a carbon reinforced carbon composite material: filament wound carbon with chemical vapor deposited matrix (FW/CVD). Contrary to public belief that the carbon composite was brittle, they observed a number of qualitative and quantitative pieces of evidence, revealing the non-brittle, forgiving nature of FW/CVD. How does one explain this unusual observation?

To date, experimentalists have been working with composites with ideal fiber orientations, such as 0°, 45°, 60°, and 90°. How does the composite behave with an arbitrary fiber orientation? What if orthogonality is no longer present?

5.5. A Comparison Between Composites and Conventional Materials

Advanced filamentary composite materials possess unique characteristics, some of which are highly desirable for high performance flight vehicles. As an example, most composites have very high strength to weight ratios, which is a great advantage. Other composites, such as graphite-epoxy, have desired thermal characteristics. A designer can exercise his

control and come up with a host of desired thermal expansion coefficients, including a negative one. In short, many fiberous composites possess such attractive physical properties that they have become increasingly popular with engineers working on high-performance products, despite the high cost.

However, the composites are not without their problems which stand in their way to becoming a much widely used engineering material. One problem area is the inadequate knowledge on fracture toughness of such composites. Indeed, more work remains to be done before a wider data based and better methods will become available to the designer. Today, the designer is heavily dependent on developmental tests and semi-empirical approaches when it comes to fracture toughness, fatigue, and stress concentration considerations.

Another problem area, in the opinion of this writer, is the relatively brittle nature of the high-performance composites (graphite-epoxy, etc., but not metal-matrix composites). Following is a list of some of the approximate fracture characteristics of materials of interest. Based on the limited data given here, it appears that composites have relatively low fracture toughness or stress intensity values. This is particularly true of the light weight composites, and not necessarily so with the heavy metal-matrix composites. The latter, of course, compromises the promised

(attractive) strength-to-weight ratio a composite system is expected to possess.

Materials	σ_u (Ksi)	G_{Ic} (in-lb/in^2)	K_{Ic} (Psi$\sqrt{in}$)	Sources
Mild Steel	160	-	150,000	Tiffany & Masters
SAE4340 Steel	230	125	64,000	Faupel
17-7PH Steel (Forged)	-	-	51,300	Tiffany & Masters
18% Ni Steel	300	-	90,000	"
Ti-6A8-4V Titanium	-	-	125,000 to 175,000	Payne
Ti-B-210 VAC Titanium	-	274	-	Faupel
Dow 332 Epoxy	-	6	1600	Broutman & Krock
Scotchply Composite	-	-	3,550	Wu and Renter
CSM/Polyester	-	-	3,070	Cherry & Harrison
Tungsten/copper	-	60	-	Tetelman
Boron/6061 Aluminum	-	-	31,000 to 42,000	Hancock
6061-T6 Aluminum	-	-	50,000	Section 4.2
2024-T3 Aluminum	-	300	47,000	Faupel
7075-T6 Aluminum	-	115	23,000 to 33,000	Kaufman & Hunsicker Faupel

6. CONCLUDING REMARKS AND RECOMMENDATIONS

As a result of this one-year part-time study on the fracture toughness of advanced fiber-reinforced composite materials, a few points have become clear. It appears that filamentary composite materials, such as boron-epoxy and graphite-epoxy, will be utilized for aerospace applications on an ever increasing scale simply because of their attractive mechanical properties. However, the fracture toughness of such materials is not well understood. To pave the way for successful future applications, it is felt that a carefully-planned effort is needed.

Among others, this effort should involve both laboratory work and theoretical investigations. In particular, the potentially useful graphite/epoxy, or any other desired composites, should be extensively fracture-tested to determine its fracture toughness and other crack propagation characteristics. Furthermore, not only unidirectionally reinforced specimens should be tested, but also those angle-plied, as fiber orientation stands out as the single most important factor influencing fracture toughness of a composite. Results obtained from these tests, of course, will also prove very useful in guiding the theoretical development, leading to a better understanding of the fracture toughness of some composite materials of interest.

In this age of modern computers, it goes without saying that computers should prove very useful in a theoretical investigation into the fracture of composites. A few recent publications dealing with computer-numerical evaluation of fracture toughness of conventional materials, appear to have laid down a good foundation for researching into composite fracture areas.

It is noted with interest that the Charpy impact test is relatively easy to perform. In the past 30 years, it has been widely used to provide useful information regarding the fracture behavior of homogeneous, isotropic, metallic materials. The desirability of using results from Charpy impact test, in lieu of fracture toughness or stress intensity factors which can be obtained in a conventional fracture toughness test, appears worth investigating.

As is true in structures made of conventional materials, stress concentrations are expected to exist in composite materials. Further, the phenomenon of stress concentration is closely related to the fracture phenomenon. Unfortunately, little is known about the nature and magnitude of stress concentrations in structural parts made of composites. Clearly, adequate efforts must be made to properly address this important problem area. It is believed that computer solutions will become feasible following a moderate effort in

testing work.

At present, the state-of-the-art in fracture toughness of filamentary composites remains at a level, approximately, described in Sections 4 and 5. It is, however, moving ahead rapidly. Nevertheless, the field is widely open with a lot to be done, particularly in the area of test data and other experimental observations related to composite fracture mechanism.

Based on results of a limited number of tests, it was suggested in Sections 4 and 5, that the linear elastic fracture mechanics based on the Griffith theory could be extended to the orthotropic materials of which the fiber composite material is one. At this time, most evidences available do support this suggestion, despite the question of heterogeneity of composites. Whether this tentative conclusion is also valid with the general filamentary composites remains to be seen.

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13. ABSTRACT

The basic concept of structural fracture in homogeneous, isotropic materials is discussed. Fundamental relationships associated with the linear elastic fracture mechanics, based on the Griffith-Irwin approach, are presented. Discussed next are properties of fiber reinforced (or filamentary) composite materials from fracture mechanics point of view. Representative fracture test results are then reviewed and analyzed. Involved in the tests are numerous composites such as beryllium/aluminum, glass/epoxy, Scotchply, boron/aluminum, graphite/polyimide, etc. Some of these nonhomogeneous anisotropic composites are unidirectionally reinforced, while other are angle-plied. Finally, the fracture phenomenon and process in filamentary composites are discussed in details. It is found that the linear elastic fracture mechanics is applicable to most composite materials which have been investigated. It is nevertheless premature and risky to predict the applicability of the theory to the filamentary composite materials in general. Also made in this report are recommendations for future work on fracture of structural composites intended for high-performance flight vehicles.

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